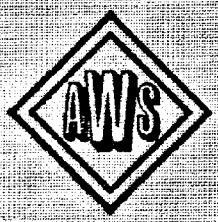
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Welding of Cast Iron

A Selection of Papers



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Welding of Cast Iron

A collection of outstanding articles from periodicals and reference literature

Prepared by the AWS Committee on Welding of Iron Castings

AMERICAN WELDING SOCIETY
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Preface

The welding of cast iron is a subject of concern to many people both in and out of the welding community. The level of confidence in obtaining dependable welds is low, particularly among those who do not have ready access to reliable information on this subject. In reality, the picture is not as dark as it may seem.

In 1965, the American Welding Society, through its D11 Committee on Welding Iron Castings, published a comprehensive study entitled "A Test Program on Welding Iron Castings". Additionally, hundreds of articles and reports have appeared in technical publications in most of the industrialized nations of the world. To this body of knowledge must be added the data contributed by many foundries and filler metal manufacturers whose research programs have developed many new applications and materials.

The AWS D11 Committee, which includes members from AWS, ASM, ASME, the Ductile Iron Society, the Iron Casting Society, and the Centre Technique des Industries de la Fonderie (France), addressed the task of reviewing nearly four hundred articles from all sources. Forty-three articles were chosen for reproduction in this reference book. The aim was to cover all types of iron castings and all commonly used welding processes. The criteria for selection were (1) completeness of information, (2) accuracy and agreement with currently accepted practice, and (3) sound background of historical fact.

The book includes a Subject Index which is cross-referenced so that pertinent information may be found under several headings.

Additional articles and reports have been listed in a Bibliography to assist the reader who wishes to delve further into a subject.

Because cast iron is nonhomogeneous and has high carbon content, welds on iron castings will not yield good results so consistently as welds on mild steel. A shop may occasionally encounter an unweldable casting. Also, a weld in an iron casting will seldom, if ever, have properties equal to those of the base metal. Nevertheless, there are many applications where useful welds can be made in iron casting when information presented here is used in conjunction with sound engineering judgement.

The Committee welcomes suggestions, comments, and constructive criticism pertaining to this publication. Please address correspondence to Secretary, D11 Committee on Welding Iron Castings, American Welding Society, P.O. Box 351040, Miami, FL 33135.

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WELDING GRAY IRON WITH MILD STEEL ELECTRODES

By R. L. KUMAR
Indian Institute of Technology
Kharagpur, India



Microstructure at the weld area shows that fusion is complete at the junction of steel (top) and gray iron.

For a listing of available free literature on welding, turn to page 133.

not in this book

IN BRIEF: Gray iron can be welded most economically by arc welding with mild steel electrodes so long as certain precautions are taken before, during, and after welding. Some key precautions are proper joint preparation, choice of electrode, preheating, and post heating. The author explains what to do and why.

■ WELDING can be a valuable tool for the gray iron foundry, not only for repairing surface defects on castings, but also for welding together two or more castings. And with welding, castings can be incorporated in assemblies including rolled, forged, or other cold-worked materials.

Under these conditions, the metallic arc welding process is the most satisfactory and economical, particularly when mild steel electrodes are used. For one thing, they are the least expensive of the available electrodes for welding gray iron. For another, the mild steel weld metal has a high solidification rate. This fact makes possible arc welding in any position, which is not possible with other welding methods because of the high fluidity of gray iron.

Success in welding gray iron with mild steel electrodes, however, requires an understanding of the process.

Metallurgical Aspects—Properties of the final component depend to a large extent on the composition and structure of the metal at the interface between the different work pieces joined together; so metallographic techniques are important in studying the structures of the gray iron, the weld metal deposited from the electrode, and the heat-affected zone at the interface. Proper understanding of the changes in volume that take place near the interface during welding also are essential.

During welding, the highly localized heat input results in steep temperature gradients between the weld metal and the cold gray iron. They produce high expansion and contraction stresses in the restricted space. Because steel shrinks more than gray iron, precautions must be taken to prevent formation of shrinkage cracks, and lack of pressure tightness can result.

Both the graphitic or free carbon and the combined carbon in gray iron go into solution in the molten metal. Simultaneously, the mild steel weld metal absorbs carbon from the molten iron, and some of the remaining carbon and silicon is burned out by the high heat at the weld zone.

Then the molten metal is cooled rapidly by the surrounding mass of cold gray iron. The rapidly cooled gray iron first solidifies as a mixture of austenite and cementite. When it cools below 750 C (about 1,380 F) the austenite changes to pearlite. At room temperature, the alloy consists of a mixture of pearlite and cementite because most of

Source: Foundry, January 1968

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the free carbon is retained in combined form. This metal, which is brittle, unmachinable white iron, is formed in the parent metal next to the interface.

On the other side of the interface, the mild steel changes to martensite under similar heating and cooling conditions. Away from the interface, the weld metal is high carbon steel. Both of those constituents impair the machinability of the weld joint.

Thus the interface area is a complex structure that includes free graphite, martensite, pearlite, and some ferrite. It presents problems if machining is needed, and it can be the source of poor strength.

The hard white iron in the weld zone is weaker than the gray iron behind it on one side of the interface, as well as the high carbon steel on the other side, and the steel weld metal behind that. When a welded joint fails in tension, it fails in the white iron area.

Usually, on gray iron castings, the welds are not required to be machined. Then, welding with the mild steel electrodes without further treatment is quite satisfactory, particularly if the strength requirements are not high and the structure of the weld metal is not important.

Recommendations — As we've mentioned, definite precautions must be taken when the common coated mild steel electrodes are used because they otherwise tend to produce cracks in the weld or at the interface.

To minimize the development of stresses, the gray iron should be heated to the bare minimum. Also, because mild steel gives easy fusion with minimum penetration, a small-gage electrode with the lowest current setting that gives sufficient penetration should be used. Although steel rods less than ⅛ in. in diam can be used, a ⅛-in. rod usually is employed with a setting of 70 to 90 amp.

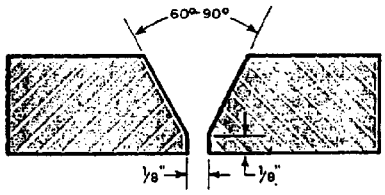
Suitable fluxes are provided as coatings on the electrodes. A typical coating contains 40 to 60% graphite, 1% barium chloride, and remainder silicon carbide.

For welding gray iron, alternating current is smoother in operation at low current values, and it produces a less penetrating arc. Reverse-polarity direct current also can be used.

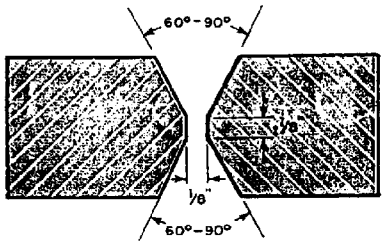
Grooves required for welding may be prepared by chipping, grinding, or cutting with a gas torch.

A single V-groove is used for metal sections up to 1½ in. thick, and a double V is made for those greater than 1½ in. thick. The V-grooves should have an included angle of 60 to 90°.

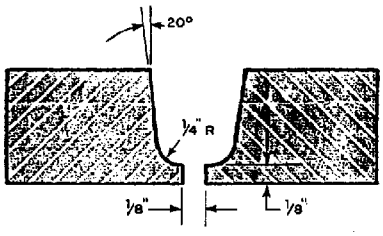
For heavy castings, a U-shaped groove with the walls of the U sloping by about 20° can be provided. This groove is intermediate between the single and the double V-groove, and welding is done from one side only.



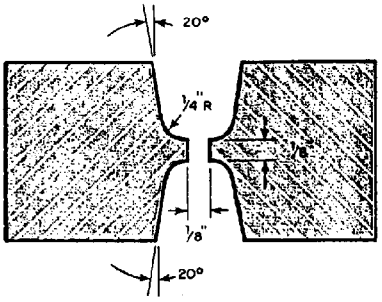
SINGLE V GROOVE



DOUBLE V GROOVE

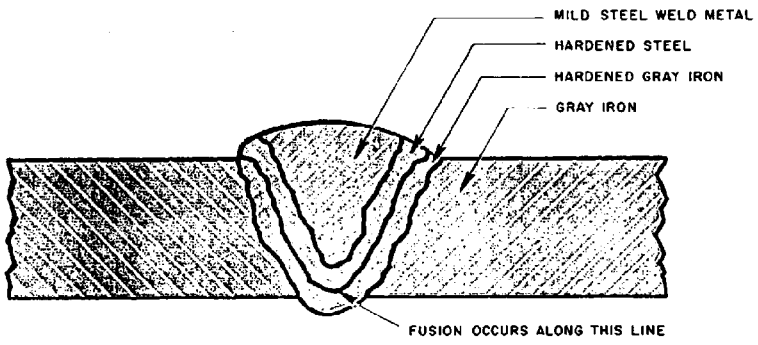


SINGLE U GROOVE



DOUBLE U GROOVE

The grooves that should be used in joining two pieces of cast iron depend on the section thicknesses involved.

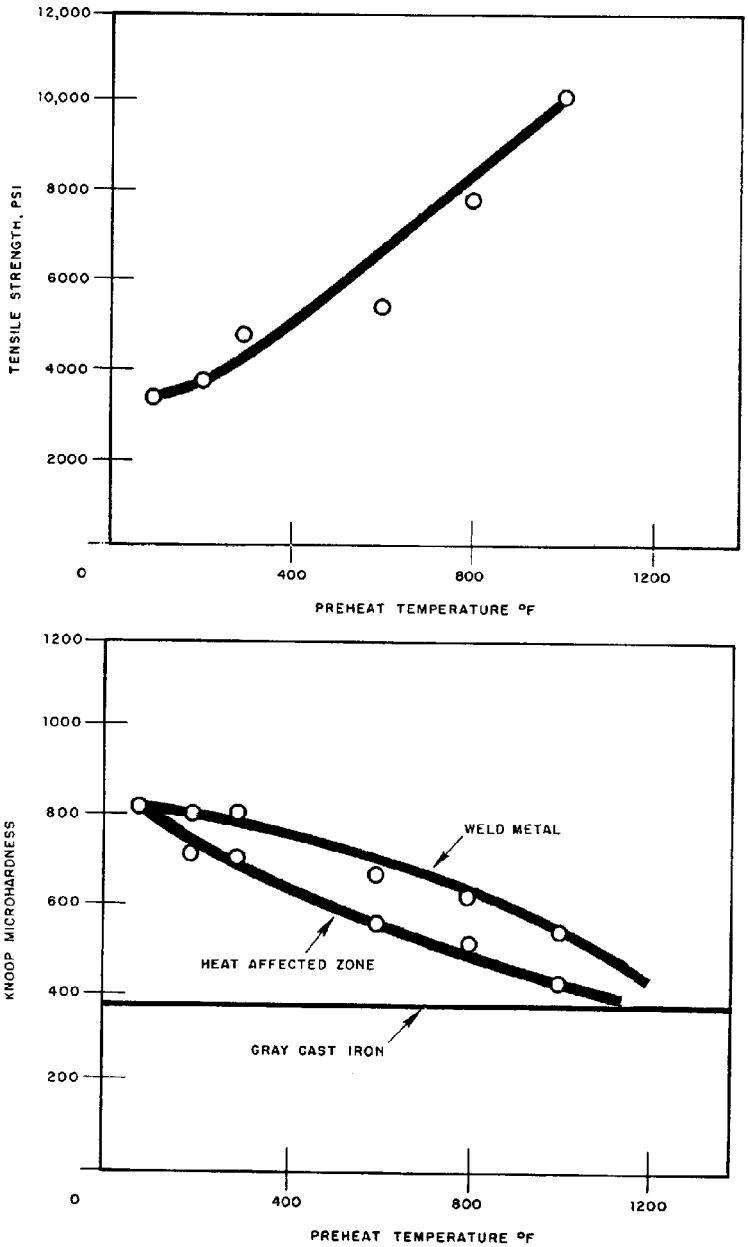


Cross-section represents the heat-affected zone when gray iron has been welded with mild steel electrodes.

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On small castings, the V-groove is made with a grinding wheel. For those castings having sections greater than 1 in. thick, a cutting torch should be used. A pneumatic chipping hammer also can be used to rough out the grooves, which then are finished by grinding. The disadvantage with chipping is that it may cover up certain defects in the castings that are critical.

Since steel contracts more than gray iron during cooling, the former tries to pull away from the latter. This effect is more pronounced when long weld beads are deposited. As the first step in overcoming this difficulty, the beads are deposited in a curved path instead of a straight line. In addition, the beads are placed in short lengths widely separated from each other so that each bead can cool for 3 to 5 min to minimize the heat absorbed by the work at one time. As a final step, peening the hot beads helps to produce good joints.



Plots relate preheat temperature and weld area microhardness.

Heat Treatment—In case machining of the weld is essential, the main problem is to obtain softer welds. Often adequate strength in the weld joint also is desirable. The best method available under such circumstances is to preheat the work to an optimum temperature of 300 C (570 F) with a maximum of 600 C (1,110 F).

Preheating reduces the cooling rate and hence prevents hardening and cracking in the weld zone. Those conditions are inevitable if the cooling rate is greater than 600 F per min. Preheating is recommended when the components to be welded are such that the expansion caused by the heat of welding results in excessive stresses in other parts of the casting.

Preheating cannot be used, however, when minimum heat is to be applied to avoid distortion of the parts being welded.

Preheating offers other problems than distortion. Even after an annealing treatment, an unmachinable area of high carbon steel remains in the weld zone, and high preheat temperatures — say above 450 C (840 F)—are likely to weaken the gray iron. Owing to the low total input, supplementary heating very often is required to maintain the preheating temperature.

The machinability of welded gray iron can be increased by a post welding heat treatment up to 650 C (1,200 F). That treatment specifically reduces the hardness of the heat-affected zone. A further reduction of hardness can be accomplished by applying a full anneal treatment that decomposes the unstable cementite to graphite and ferrite. Unfortunately, it also weakens the gray iron.

In both the above methods, local heating of the joints or total heating of welded assemblies is done at temperatures of 540 to 650 C (1,000 to 1,200 F). The castings are covered with materials like asbestos or sand that are designed to promote slow cooling.

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Weldability of Gray Iron Using Fluxless Gray Iron Electrodes for SMAW

Welds having original base metal hardness and graphite content are attained by decreasing weld cooling rate and increasing the carbon equivalent of the weld admixture

BY J. H. DEVLETIAN

ABSTRACT. The weldability of class 40 gray iron using fluxless gray iron electrodes was investigated. Single and multiple-pass welds were deposited by the SMAW process with preheating temperatures up to 830C (1550 F). The cooling rate for each weld was recorded. Evaluation of weldability included microhardness studies, metallurgical analysis and crack susceptibility of the weld and heat-affected zone (HAZ).

The weldability of gray iron was significantly affected by the cooling rate and carbon equivalent of the weld admixture. Decreasing cooling rate and/or increasing carbon equivalent substantially lowered weld hardness due to the increased amounts of graphite and ferrite forming in the weld. Peak HAZ hardness of the gray iron base metal was also found to decrease with decreasing weld cooling rate.

Cracking tendencies in the weld and HAZ were eliminated by maintaining the cooling rate below 10 C/s or 1.8 F/s (or above 300 C or 572 F preheat). Crack-free welds having original base metal hardness and graphite content were readily attained by reducing the cooling rate (through preheating) to a level that was dependent upon the carbon equivalent of the weld admixture.

Introduction

The utilization of inexpensive gray iron electrodes to weld or repair-weld large gray iron castings by the shielded-metal arc welding (SMAW) process has been largely ignored in favor of the more costly nickel-base electrodes—namely ENi-CI and ENiFe-CI. Despite the increased cost and the consumption of the strategic element nickel, many foundries have been reluctant to use gray iron electrodes for fear of producing cracked white iron weld metal with a martensitic heat-affected zone (HAZ).

The gray iron literature is replete with research concerning the factors affecting the degree of graphitization of gray iron castings and its associated mechanical properties. It is well-known that high quality gray iron castings can be obtained by:

1. Slow cooling rates during eutectic and eutectoid transformations.
2. Carbon equivalent (C.E.) of near-eutectic composition.[1]
3. Use of graphitizing inoculants.[2,3]
4. Low superheating temperatures.[4,5]

Under these conditions the resulting microstructure will contain the desired graphite flakes in a matrix of ferrite and/or pearlite.

The welding of gray cast iron with gray iron electrodes, however, is particularly difficult because the effective use of inoculants and low superheating temperatures are not practical in the SMAW process. Successful welding of gray iron would require a sufficiently slow weld cooling rate and a near-eutectic C.E. in the weld admixture to insure a high graphitizing potential; otherwise, a white iron weld is inevitable. Recently, Hogaboom[6] reported that fluxless cast iron electrodes could be used to repair-weld large gray iron castings provided that adequate preheating, high welding currents and controlled cooling proce-

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Table 1—Chemical Composition of Gray Iron Electrode and Base Metal, %

	C	Si	P	Mn	S	C.E.[a]
Base metal	2.87	1.80	0.10	0.81	0.067	3.4
Electrode	3.84	2.92	0.35	0.75	0.020	4.5

[a]C.E. = %C + .3 (%Si + %P)

Table 2—Chemical Compositions of Weld Admixtures Deposited on Gray Iron, %

	C	Si	P	Mn	S	C.E.
Single pass weld	3.00	2.30	0.20	0.69	0.040	3.8
Multiple pass weld (Analysis on final pass)	3.35	2.66	0.29	0.70	0.015	4.2

dures were utilized.

The purpose of the study reported here was to further refine the existing work on the weldability of gray iron using fluxless gray iron electrodes and the SMAW process. Specifically, this paper quantitatively relates the microhardness, degree of graphitization, microstructure and crack susceptibility of the weld admixture as a function of measured values of weld cooling rate and C.E.

Procedure

Materials

ASTM class 40 gray cast iron in the as-cast condition was used exclusively in this investigation. Ingots approximately 20 × 20 × 70 cm (7⅞ × 7⅞ × 27⁹⁄₁₆ in.) were machined into 10 × 10 × 2 cm (3¹⁵⁄₁₆ × 3¹⁵⁄₁₆ × ¹³⁄₁₆ in.) test plates for subsequent welding. A 90 deg single V-groove was also machined into half of the test plates. In order to duplicate the repair-welding procedures used in commercial practice, the surfaces to be welded were cleaned by using a hand operated grinder.

Filler metals used to weld gray iron were 5 × 5 mm (0.2 × 0.2 in.) square class 20 gray iron and 3.2 mm (⅛ in.) diameter ENi-CI electrodes. Compositions of all materials are given in Table 1.

Welding

Single pass, bead-on-plate welds were deposited on the flat test plates using the manual SMAW process. Multiple-pass welds were deposited on the test plates containing the 90 deg single V-groove. The interpass temperature was either the specified preheating temperature or room temperature when no preheating was used.

All SMA welds made with fluxless gray iron electrodes were deposited at a constant arc energy input of 2700 J/mm, using a welding current of 450 A and a travel speed of approximately 2.5 mm/s (5.9 ipm). The ENi-CI electrodes used 160 A and approximately 6 mm/s travel speed.

Control of Weld Cooling Rate

A weld cooling curve was obtained for each weld specimen by plunging a 0.38 mm (0.015 in.) diameter W-3%Re/W-25%Re thermocouple directly into the molten weld pool. Cooling curves were recorded on a continuous chart recorder and the weld cooling rate for each weld was taken at 704 C (1300 F). Weld cooling rates ranging from 0.19 to 160 C/s (0.35 to 288 F/s) were investigated.

Slow weld cooling rates were obtained by preheating the workpiece on a heavy-duty electric hot plate. Accurate preheating temperatures were assured by welding a chromelalumel thermocouple onto each test plate. As soon as the thermocouple registered the desired preheating temperature in the test plate, a variac attached to the hot plate was adjusted to stabilize this temperature. After stabilization, the power to the hot plate was shut off and welding commenced immediately.

The fastest weld cooling rates were obtained by partially immersing the 20 mm (¹³⁄₁₆ in.) thick test plates in a shallow pool of water about 5 mm (0.2 in.) deep and then welding. The water which was maintained at 25 C (77 F) acted as an efficient heat sink during welding.

Metallography

All metallography of weld and HAZ structures were performed on transverse-to-weld sections taken within 1 cm (0.04 in.) of the embedded thermocouple for an accurate correlation between weld cooling rate and microstructure. Sections were polished and then etched in 4% picral.

Carbon Equivalent

To study the effects of C.E. on weld properties, two values of C.E. (namely, 3.8% and 4.2%) were compared throughout this investigation.

The value of 3.8% was obtained by welding class 40 gray iron having a C.E. of 3.5% with a class 20 gray iron electrode with a C.E. of 4.6%. The resulting weld metal admixture possessed the C.E. of 3.8%—Table 2.

Similarly, the 4.2% C.E. value was obtained for the multiple-pass welds where the final pass of the resulting weld admixture contained a C.E. of 4.2%—Table 2.

Microhardness

Diamond pyramid microhardness tests were conducted on transverse-to-weld sections using a 300 gram (0.66

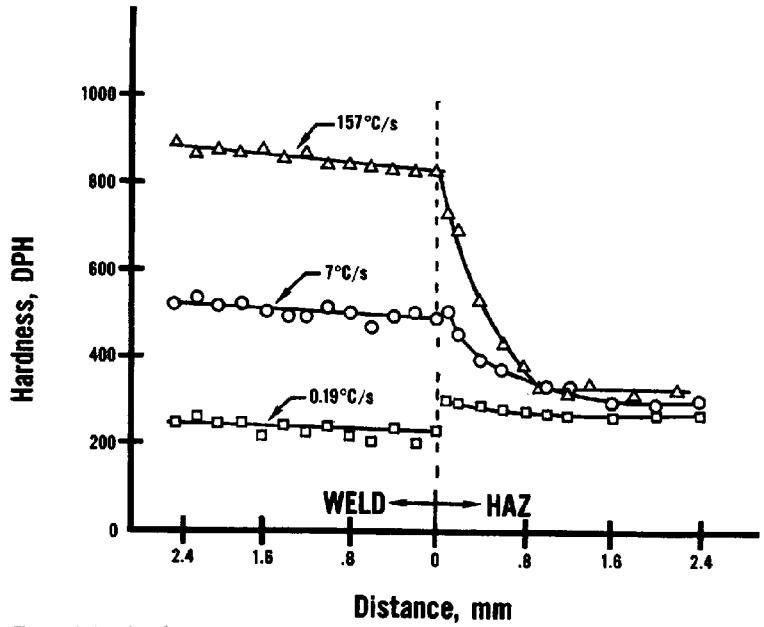


Fig. 1—Microhardness scan over the weld and HAZ for fast, intermediate and slow weld cooling rates. Weld admixture contains 3.8% C.E. Base metal is class 40 gray iron

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lb) weight acting on the indenter. The average hardness of each polished and etched weld was obtained by taking at least ten equally spaced indentations (at 0.2 mm (0.08 in.) intervals) vertically and at least 10 readings horizontally within the weld fusion zone. Similarly, the average value of the peak HAZ hardness was obtained by taking at least six indentations along the HAZ side of the weld-base metal interface. Care was taken not to indent any of the large graphite flakes in the HAZ.

The diamond pyramid microhardness test was chosen because of its ability to provide the microhardness values of any part of the weld/HAZ microstructure. This test was used to obtain the hardness of the matrix and not that of large graphite flakes. Thus, the diamond pyramid hardness values tend to be slightly higher than similar Brinell or Rockwell test results when large graphite flakes are present in the microstructure.[7]

Weld Cracking

Weld cracking was reported after visual and dye-penetrant examinations were performed on the single-pass and multiple-pass welds.

Results

The weldability of gray cast iron using gray iron filler metal and the SMAW process was evaluated by hardness tests and metallographic analysis. Results of this work are presented in Figs. 1-8.

Hardness

The hardness profile across the weld and HAZ of a single-pass (C.E. = 3.8%) weld was found to be dependent upon weld cooling rate. Typical microhardness scans for welds having fast, intermediate, and slow cooling rates are shown in Fig. 1. The fast cooling weld resulted in the highest overall hardness values in the weld and HAZ, while the slow cooling weld resulted in lowest hardness.

Average weld metal hardness values for both the 3.8% and 4.2% C.E. welds were found to decrease with decreasing weld cooling rate as shown in Fig. 2a. The hardness values for the 4.2% C.E. welds were consistently lower than those for the 3.8% C.E. welds. For example, at a weld cooling rate of 3 C/s (5.4 F/s) , the average hardness values for the 3.8% and 4.2% C.E. welds were 420 and 265 DPH, respectively.

Weld metal hardness values were comparable with those of the gray iron base metal only when the weld cooling rates for the 3.8% and 4.2% C.E. welds were reduced to approximately 0.7 and 5.3 C/s (1.3 and 9.5 F/s), respectively. By contrast, all welds deposited with the ENi-CI electrodes (for reference only) were much softer than the gray iron base metal regardless of weld cooling rate.

The peak HAZ hardness of the gray iron workpiece was also affected by weld cooling rate. In Fig. 2b, peak HAZ hardness decreased significantly with decreasing weld cooling rate.

Metallographic Analysis

Graphite Formation. In this study, both the amount and size of the graphite flakes which resulted during eutectic solidification were found to increase substantially with decreasing values of weld cooling rate and

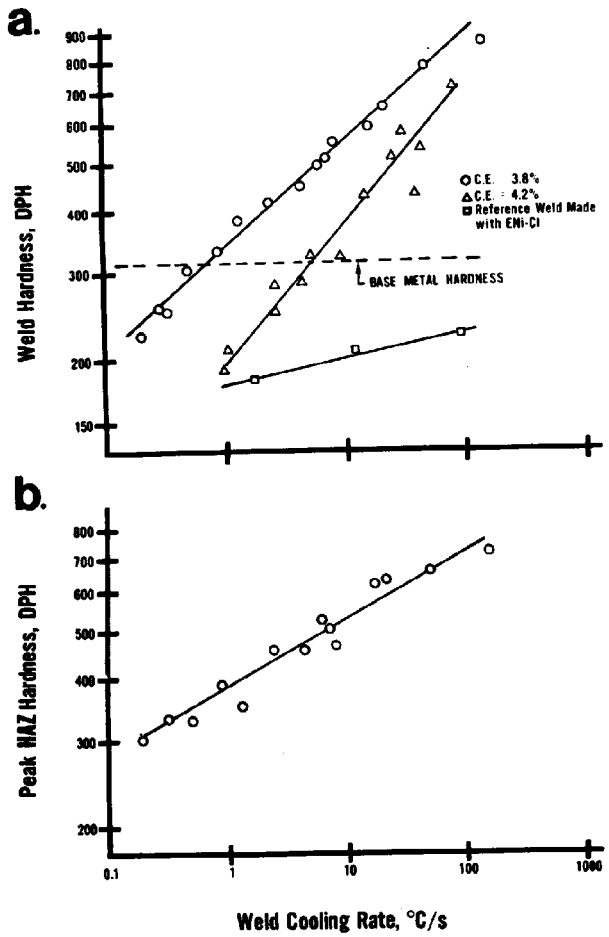


Fig. 2—The effect of weld cooling rate on: a—average weld metal hardness for admixture compositions of 3.8% and 4.2% C.E.; b—peak HAZ hardness of gray iron

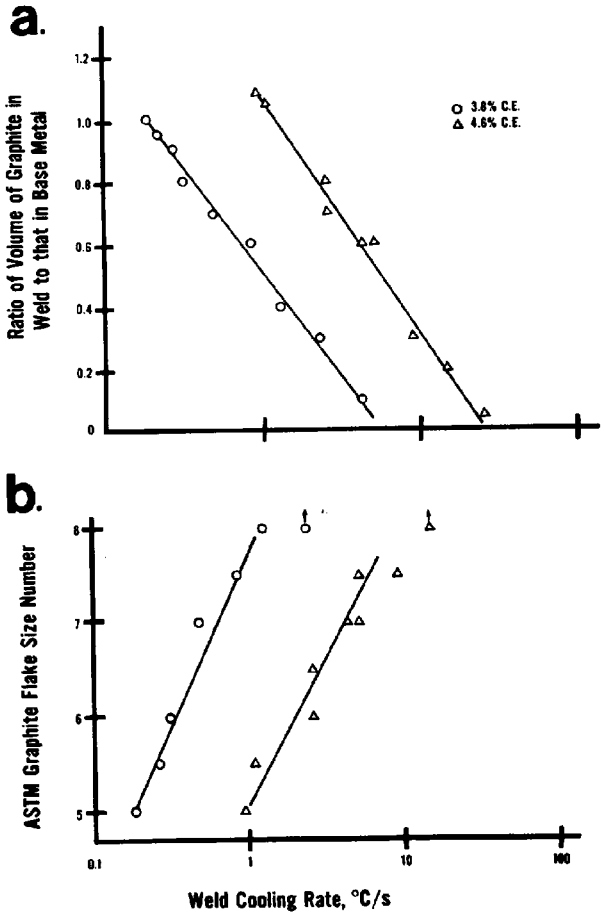
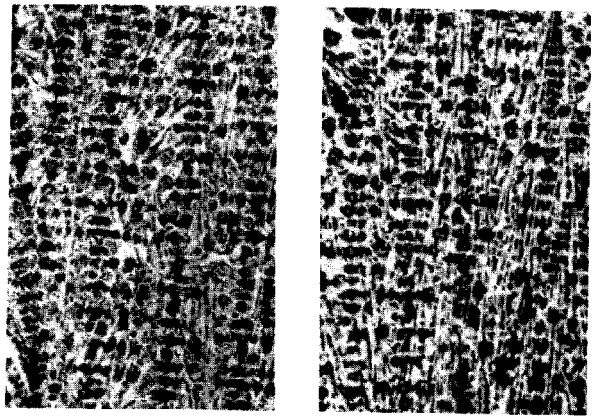


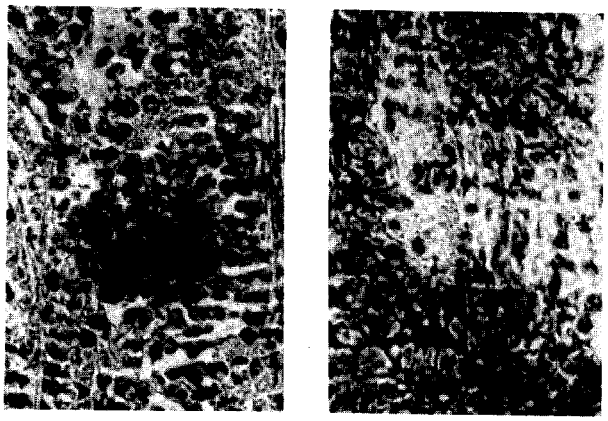
Fig. 3—The effect of weld cooling rate on: a amount and b size of graphite flakes forming in 3.8% and 4.2% C.E. weld metal

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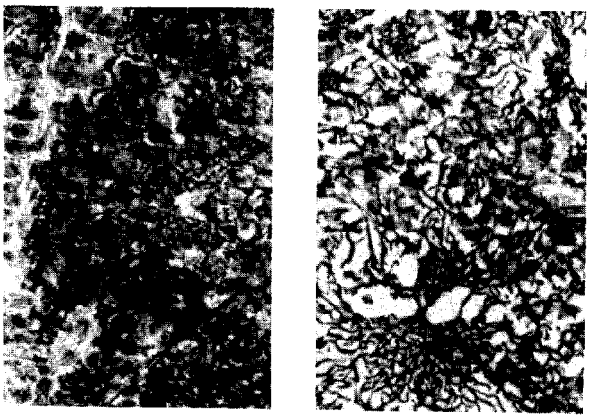
a b

Fig. 4—Rapidly cooled welds deposited on gray iron having been preheated to 200 C (392 F). Weld in **a** *has a 3.8% C.E. and 22 C/s (39.6 F/s) cooling rate while the weld in* **b** *has a 4.2% C.E. and a 26 C/s (46.8 F/s) cooling rate. Picral etchant; ×400 (reduced 50% on reproduction)*



a b

Fig. 5—Welds having an intermediate cooling rate were preheated to 400 C (752 F). Weld in **a** *has a 3.8% C.E. and 4.2 C/s (7.6 F/s) cooling rate while the weld in* **b** *has a 4.2% C.E. and 5.2 C/s (9.4 F/s) cooling rate. Picral etchant; ×400 (reduced 50% on reproduction)*



a b

Fig. 6—Slowly cooled welds deposited on gray iron having been preheated to 590 C (1094 F). Weld in **a** *has a 3.8% C.E. and 0.89 C/s (1.6 F/s) cooling rate while the weld in* **b** *has a 4.2% C.E. and 0.95 C/s (1.7 F/s) cooling rate. Picral etchant; ×400 (reduced 50% on reproduction)*

increasing C.E.—Fig. 3. The amount and size of graphite flakes observed in the 4.2% C.E. welds were consistently greater than those observed in the 3.8% C.E. welds. Generally, the form of graphite flakes observed in all welds were predominantly the interdendritic ASTM types D and E.[8] No graphite was observed in 3.8% C.E. and 4.2% C.E. welds for weld cooling rates greater than approximately 6 and 30 C/s (11 and 54 F/s), respectively.

The weld microstructures in Figs. 4-7 illustrate the effect of weld cooling rate and C.E. on the formation of graphite. The fast cooling welds in Fig. 4 are clearly white iron. In Fig. 5, intermediate cooling conditions resulted in a mottled iron structure with the 4.2% C.E. weld containing significantly more graphite than the 3.8% C.E. weld. The microstructure of the slowly cooled welds in Fig. 6 shows that the 4.2% C.E. weld was fully gray iron while that of the 3.8% C.E. weld was still mottled iron. It was not until the weld cooling rate was reduced to 0.32 C/s that a fully gray iron structure developed in the 3.8% C.E. weld as shown in Fig. 7.

Solid State Transformations. Once the eutectic structure had solidified, the subsequent eutectoid transformation proceeded very much like that of a high carbon steel. For the fast cooling welds, both the primary dendrites of austenite and the austenite in ledeburite transformed to a bainitic structure—Fig. 4.

With decreasing cooling rate, this bainitic structure was replaced by fine pearlite—Fig. 5. At very slow cooling rates, the solid state transformation of austenite resulted in pearlite and ferrite as shown in Figs. 6 and 7.

Cracking in Welds

Severe transverse cracks resulted in both the 3.8% and 4.2% C.E. welds when the cooling rate was excessive, while no cracking occurred for the slowly cooled welds. In this investigation, cracking tendencies in the weld and HAZ were eliminated by reducing the weld cooling rate below approximately 10 and 36 C/s (18 and 65 F/s) for the 3.8% and 4.2% C.E. welds, respectively.

Preheating Temperature vs. Weld Cooling Rate

Following the preheating procedures outlined in this study, the preheating temperature was found to be a consistently workable control over weld cooling rate. With welding parameters fixed for all welds, increasing the preheating temperature decreased the weld cooling rate as shown in Fig. 8.

Discussion

The two most significant variables affecting the weldability of ASTM class 40 gray iron using class 20 gray iron electrodes were found to be the cooling rate and the C.E. of the weld metal. These two factors had a pronounced effect on the weld hardness, formation

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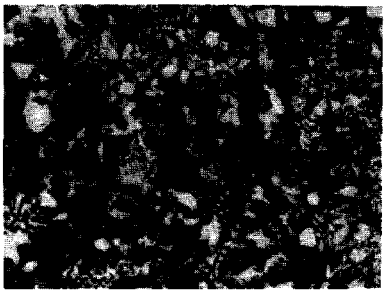


Fig. 7—Very slowly cooled weld deposited on gray iron having been preheated to 760 C (1400 F). This weld has a 3.8% C.E. and 0.32 C/s (0.58 F/s) cooling rate. Picral etchant; ×400 (reduced 50% on reproduction)

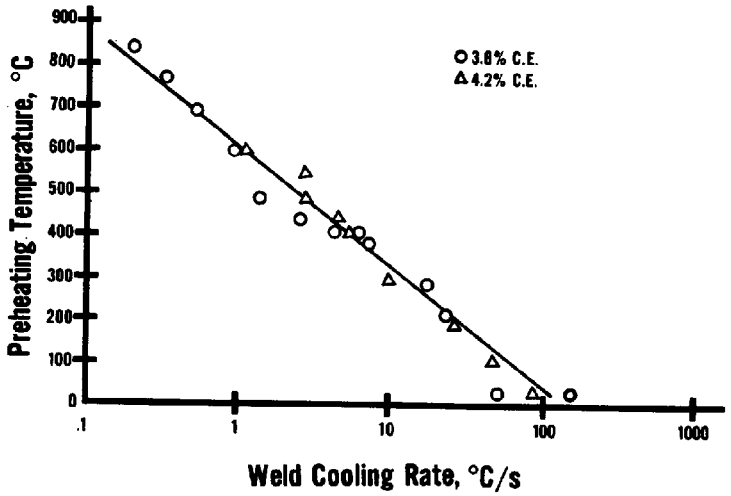


Fig. 8—The effect of preheating temperature on the cooling rate of single-pass gray iron welds where the arc-energy input, specimen size and geometry were held constant

of the Fe-graphite eutectic, eutectoid transformation and weld cracking.

Hardness and Carbon Equivalent

In this investigation, the hardness of cast iron weld metal of a given C.E. was found to depend directly upon the weld cooling rate—Fig. 2. This behavior is in agreement with the cast iron literature[1,7,9,10] in that:

1. The higher temperature eutectic reaction of austenite + graphite dominates during solidification of slowly cooled (i.e., with minor undercooling) castings.
2. The lower temperature eutectic reaction of austenite + carbide dominates for fast cooling rates (i.e., with large undercooling) resulting in white iron.

In this investigation, the occurrence of gray, mottled or white iron in the weld admixture of a given C.E. depended essentially upon weld cooling rate.

The C.E. of the weld admixture also had a significant effect upon weld hardness. Since increasing the silicon and carbon contents intensifies the graphitization potential of the weld metal, the 3.8% C.E. welds developed significantly higher hardness and less graphite than the 4.2% C.E. welds.

Microstructure

Since the formation of eutectic graphite depends upon nucleation and growth kinetics,[10] the amount and size of the flakes are controlled by the solidification time and the graphitization potential of the weld admixture. In welding of gray iron, the amount and size of the graphite flakes in the weld admixture increased with decreasing weld cooling rate and increasing C.E.

Because graphite-nucleating inoculations and low superheating temperatures are not practical for welding and, therefore, not used in this study, the morphology of the eutectic graphite forming in cast iron weld metal is predominantly the interdendritic ASTM types D and E (see Figs. 5 and 6a). It is only at extremely slow weld cooling rates that some non-interdendritic types of graphite are observed as in Figs. 6b and 7.

Furthermore, the eutectoid transformation also contributed to the resulting hardness of the weld metal. The slower the weld cooling rate and the greater the C.E., the more potent are both the solid state graphitizing and ferritizing effects. For example, at the slow cooling rates indicated in Figs. 6 and 7, solid state graphitization has permitted the formation of some ferrite during the eutectoid transformation. This is why the ferritic areas are usually located adjacent to graphite flakes.

Increased C.E. substantially augments the graphitizing and ferritizing potentials of the weld metal as can be seen by comparing Figs. 6a and 6b where the ferrite content of the 4.2% C.E. weld is far greater than that of the 3.8% C.E. weld.

Preheating Welds

The preheating temperature required to obtain a soft gray iron weld and HAZ depends upon several factors:

1. C.E. of the weld admixture.
2. Section size and configuration of workpiece.
3. Arc-energy input.

A high C.E. significantly increases the graphitizing and ferritizing potentials of the weld. Thus, increasing the weld C.E. would allow a lower preheating temperature (or faster weld cooling rate) to produce a specified weld hardness value. For example, to obtain a weld hardness of 325 DPH (see Figs. 2a and 8), the 3.8% C.E. weld requires a preheating temperature of 600 C or 1112 F (or a cooling rate of 0.88 C/s, i.e., 1.6 F/s) while the 4.2% C.E. weld needs to be preheated to only 360 C or 680 F (or 6.3 C/s i.e., 11.3 F/s). This would represent a substantial cost and energy savings.

In this study, the size and shape of the workpiece and the arc-energy input were held constant. Under these restricted conditions, the relationship between the preheating temperature and the resulting weld cooling rate is shown in Fig. 8. However, if the section size were reduced or the arc-energy input were increased, the curve in Fig. 8 would shift to the left—i.e., lower preheating temperatures to obtain a given value of weld cooling rate. In fact, Hogaboom[6] reported that SMA repair welds made on gray iron could be deposited with such high values of arc-energy input that no preheating at all was required.

The results of this study show that, metallurgically, there is no upper limit for preheating gray iron. The higher the preheating temperature, the greater the degree of graphitization and ferritization which produces softer weld and HAZ structures. However, an upper limit may be required only when excessive softening can not be tolerated.

Conclusions

The effects of SMA welding of ASTM class 40 gray cast iron using gray iron electrodes were evaluated by microhardness and metallographic analysis. Based on this study, the following conclusions were reached:

1. Weld metal and peak HAZ hardness decreases significantly with decreasing weld cooling rate.

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2. Increasing the C.E. of the weld admixture decreases weld hardness and reduces the need and cost for preheating. E.g., the 4.2% C.E. welds develop original base metal hardness and graphite content with a weld cooling rate of 5.3 C/s or 9.5 F/s (or 390 C or 734 F preheat); but decreasing the C.E. to 3.8% requires a substantially slower cooling rate of 0.7 C/s or 1.3 F/s (or 640 C or 1184 F preheat) to attain similar results.

3. The eutectic graphite content of the weld admixture increases with decreasing cooling rate and/or increasing carbon equivalent, but its morphology is primarily the interdendritic ASTM types D and E. No graphite occurs in the 3.8% and 4.2% C.E. welds when the weld cooling rates exceed 6 and 30 C/s (10.8 and 54 F/s), respectively.

4. Weld cracking due to the formation of brittle white iron is highly probable when the weld cooling rate exceeds about 10 C/s or 18 F/s (or less than 300 C or 572 F preheating).

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WELDING OF GRAY CAST IRON

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The welding of cast iron is usually considered to be rather undependable, erratic and a sort of hit-or-miss deal. This is partly because of an apparent widespread lack of understanding of the various factors associated with the welding of cast iron—also partly because of inconvenience in welding cast iron in the more dependable ways, partly because of a scarcity of suitably skilled welders versed in the more dependable methods and partly because of a scarcity or complete absence of published information concerning the fundamentals of cast iron welding and more dependable methods.

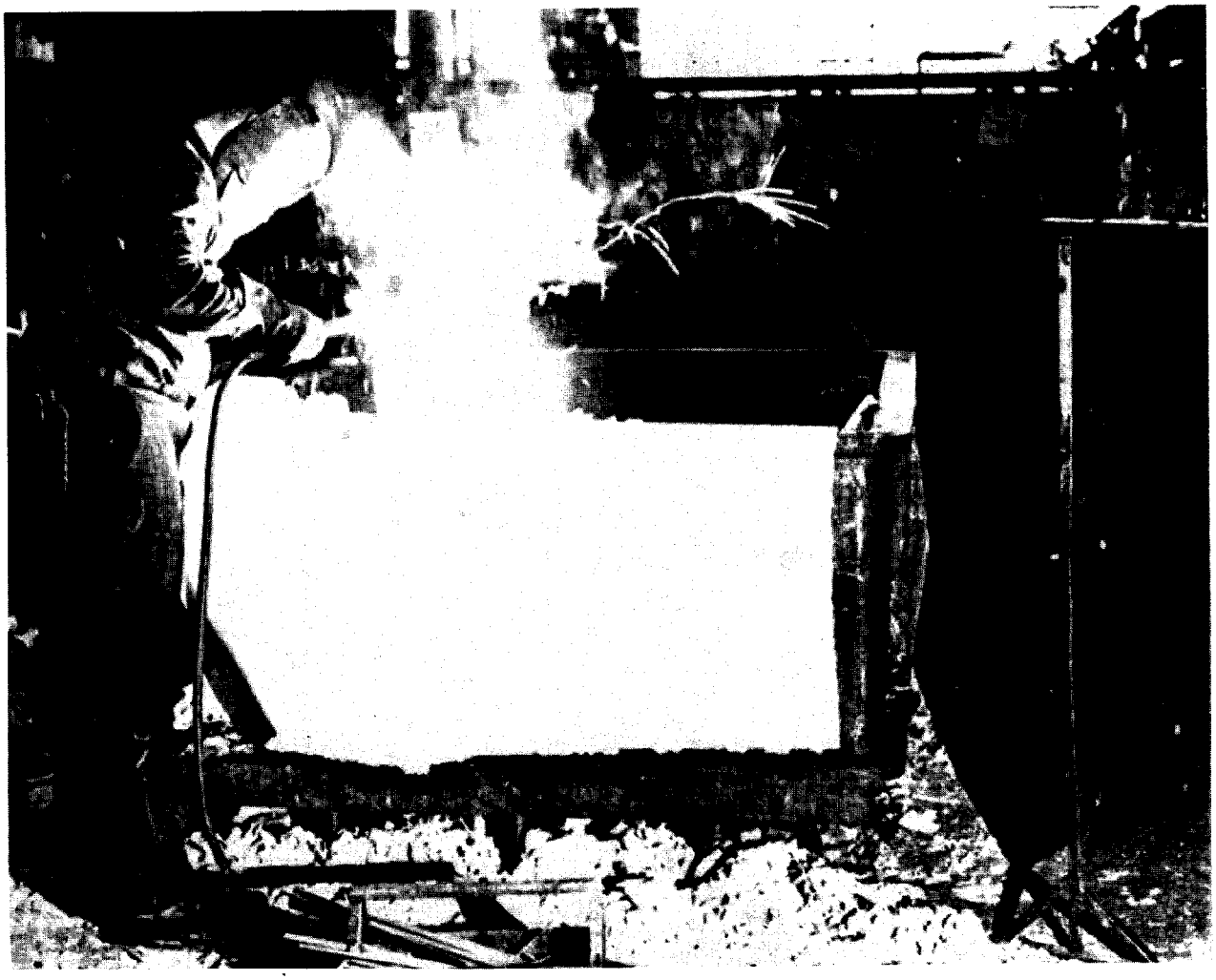
The intent of this paper is to provide at least a little insight into some of the major interrelating factors and outline a little known but very fast, economical and high quality way to make certain types of major repairs.

The Problem

Cast iron—or at least the gray variety—is a rather crude, impure alloy of mainly iron, carbon and silicon. Its formation during solidification and cooling is such that much of the carbon separates out of solution as free graphite in flake form in a matrix of rather impure, high silicon, medium to high carbon steel. The metallurgical structures typical of gray cast iron in both the unetched and etched conditions are shown in Figs. 1 and 2.

The graphite flakes act like tiny cracks within the matrix metal to promote fracture without appreciable yielding. At the same time, the high carbon, silicon and impurity content of the matrix material tends to limit its ductility and ability to deform without fracture. Combined, the two factors of crack-like discontinuities

"Hot Arc" welding in progress on a gray cast iron casting



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Fig. 1—Typical unetched structure of gray cast iron showing graphite flakes. ×100



Fig. 2—Etched structure of gray cast iron-mixture of ferrite pearlite and graphite flakes. Nital etch, × 100

and limited ductility of the matrix material result in the no ductility characteristics of gray cast iron.

Graphite is black, and the many flakes exposed on fracture surfaces give gray cast iron its characteristic gray appearance.

Hardening Characteristic

Alloys of iron and carbon, especially when accompanied by other alloying elements such as manganese, silicon, chromium, nickel, etc., are quite readily hardened by rapid cooling from high temperature. Files, knife blades, springs, the gears and axles in automobiles and many other things are hardened and strengthened in that way. However, they are usually given a second, lower temperature treatment to partly soften and toughen the parts.

The area immediately adjacent to a weld pass is heated to the melting point and cools very rapidly to or nearly to adjacent metal temperature by conduction of heat into the surrounding metal. If that temperature is fairly low—say below about 500 F (260 C)—and the material is high in carbon such as gray cast iron, extreme hardening is almost certain, irrespective of the kind of filler metal used. About the only way to prevent the hardening is to heat the surrounding material to a high enough temperature to slow down the cooling rate sufficiently to prevent hardening. In other words, preheat it. The temperature usually needed to prevent severe hardening of the heat-affected zones adjacent to arc weld passes on cast iron is about 600 F (316 C) or a bit higher.

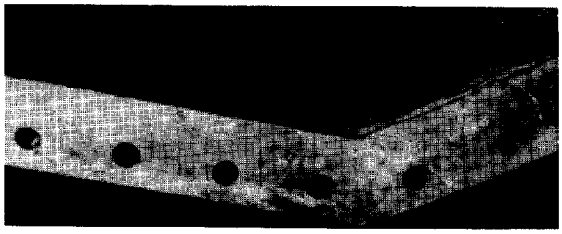


Fig. 3—Finish machined "Hot Arc" welds



Fig. 4—Preheating of gray iron casting for "Hot Arc" welding

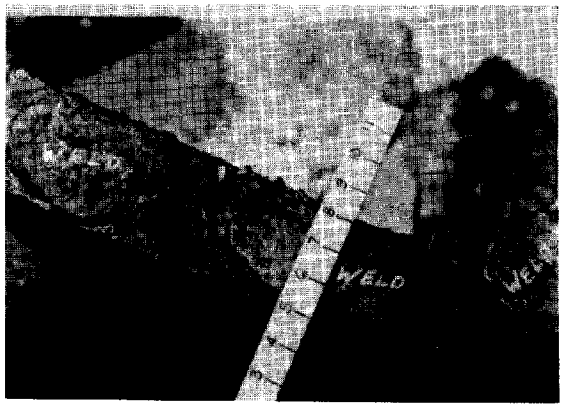


Fig. 5—"Hot Arc" weld—showing surface finish before machining

Stress Build-Up

Stresses are another important factor. Most materials expand on heating and shrink on cooling. Cast iron is no exception. On heating, the heat-affected zones adjacent to weld passes expand, soften, upset, and shrink on cooling. So also does the deposited weld metal shrink on cooling.

Shrinkage or attempted movement of anything attached to something else that is not similarly moving will set up stress—twisting a child's ear or trying to take off in a car while tied to a tree are examples of a non-metallurgical sort. For weld passes, the stress magnitude will depend on temperature difference, shrinkage characteristics of the materials involved, span over which the shrinkage difference takes place and the elastic, plastic and strength characteristics of the materials involved.

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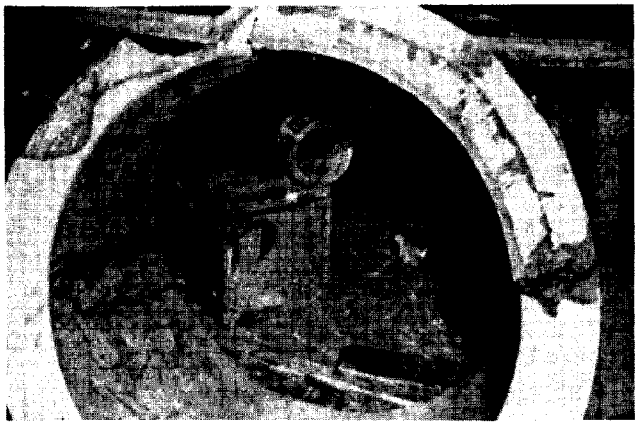


Fig. 6—Example of excavation and area damming for "Hot Arc" welding (see Fig. 7)

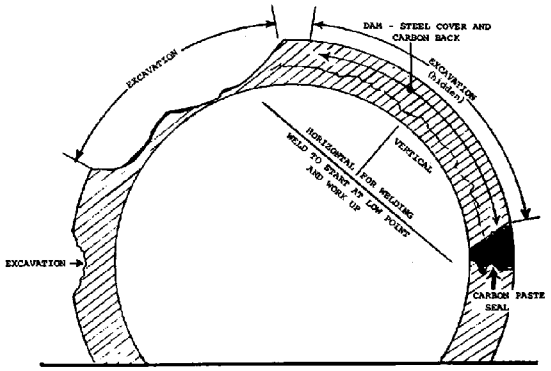


Fig. 7—Schematic representation of Fig. 6 showing excavation areas and an example of damming

About the only way to prevent stress build-up is to lessen the degree of temperature and shrinkage difference or develop easy yielding, putty-like conditions in the adjoining metals. Heating the base metal to a sufficiently high temperature—say, around 1500 F (816 C) or higher—will accomplish the latter. Heating to somewhat lower temperature—say, around 1000 F (538 C)—will lessen stress build-up sufficiently for most, if not all, practical purposes. Furthermore, heating cast iron to much over 1000 F for any extended period of time will cause considerable weakening of the base metal. All of these factors—hardening, stress and possible weakening—need to be taken into account in welding cast iron.

Basically, then, the problem boils down to near zero ductility of the base metal, tendency to harden along low preheat weld passes, stress build-up from shrinkage of passes and almost certainty of at least some cracking adjacent to welds unless sufficient preheat is used to eliminate both hardening and high stress build up. This being the case, it is in order to consider process characteristics, application problems and popular opinion concerning the welding of cast iron.

Processes for Casting Repair

Nowadays, when a gray iron casting needs repair, first thought usually turns to the possibilities of repair with nickel or nickel alloy electrodes and little or no preheat. This, of course, involves rapid heating and cooling of heat-affected zones, the hardening thereof, differential shrinkage, high stresses and a likelihood or certainty of at least some cracking in or along the welds. Consequently,

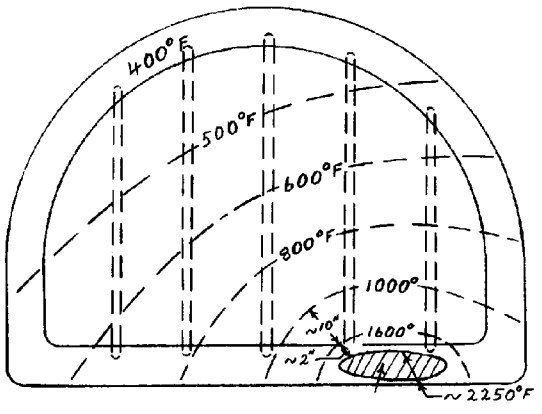


Fig. 8—Typical temperature gradient pattern for "Hot Arc" welding large cast iron parts

there is little or no possibility of producing welds in this way with over about 60 to 75% of the strength of the base metal and often much less.

If the foregoing strength and consistency are not considered adequate for the application involved, next thought would probably turn to the possibilities for gas welding with preheat, or brazing. The gas process is capable of producing welds of quite high strength and more consistently than is generally possible by arc welding with nickel alloy electrodes. However, gas welding suffers from marked inconvenience, due to the heat associated with welding on preheated parts with a big gas flame. Furthermore, the process is relatively slow and this extends the time workmen are exposed to the disagreeable heat from the gas torch and preheated base metal. Because of these factors, there are now very few men who are experienced or knowledgeable in the art of gas welding or willing to stand the heat involved in gas welding large gray iron castings. The result is usually to make the process unattractive or inapplicable, although capable of producing quite good welds.

As for brazing, this is a very good process for some repairs and is easier to apply than gas welding with cast iron rod. Bond strength may, however, be somewhat variable. Also, for high stress and cyclic load applications, the yield and fatigue strength of the brass filler metals (so-called bronzes) may also be too low.

Another method for repair of gray iron parts (but one that does not involve heat or welding) is a proprietary mechanical cross-link method called "Metalock." It is basically similar to stitching a seam with needle and thread; the process involves drilling and the insertion of tie-bars across the fracture surfaces. For some applications, to avoid dismantling, delay or distortion problems in welding, this may be a quite favorable method of repair.

If none of the foregoing methods are considered adequate or applicable for one reason or another, what other possibility (or possibilities) exist? Although relatively unknown, there is another method for welding gray iron. It is far faster than nickel electrode or gas welding and is capable of producing higher strength welds than any of the foregoing repair methods. It may be termed "Hot Arc" welding with cast iron electrodes. By this method, welds can readily be made very rapidly with the full strength of the base metal; also, the welds are both machinable and gray iron in character. Figure 3 shows the type of weld obtainable. Details of the process follow.

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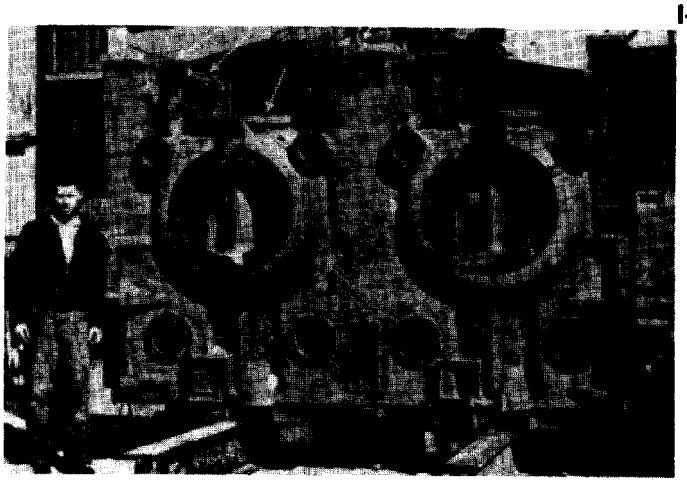


Fig. 9—Gray iron casting which had rather large shrink and sand inclusion areas repaired by "Hot Arc" welding. Arrows indicate weld regions

"Hot Arc" Process

Process Characteristics

Basically, the "Hot Arc" process is similar to gas welding except that cast iron weld metal is deposited by the metal arc process. Preheat is essential, the same as for gas welding—say, around 900 to 1100 F (482 to 593 C). The welding is then done by the arc process using cast iron rod for the electrode. The latter may be ordinary gas welding rod such as is available at any welding supply house or special rod made with a little higher carbon and silicon contents to compensate for burn-out during welding. Rods may also be coated with graphite or flux, if so desired, but coatings are not essential.

For best results, the welding, itself, should be done with rather large electrodes at high amperage and with a big, fluid puddle, rather than with quick solidifying "beads" such as are most common in welding most other materials. Typical is use of 1/4 in. diameter cast iron rod with around 400 to 450 A, or larger rod with up to 600 A.

So applied, deposit rates will run from around 10 to 30 pounds per hour (4.5 to 13.6 kg/h) and may be further increased by feeding electrically neutral filler metal (i.e., cold rod) from the side, by placing broken pieces of rod along run paths, or by feeding in cast iron chips.

The type of weld obtainable by this process and machinability thereof are shown in Fig 3. Figure 4 and the lead photograph show preheating and welding of a gray iron casting. Figure 5 shows a finished weld before machining, the same weld as shown in Fig. 3 after machining.

Area Damming

Due to the high amperage used and extreme fluidity of superheated cast iron, the hot arc process cannot be used in other than the flat position and even for that must be surrounded and supported by some sort of solid material. The root of through joints and edges of weld areas are best supported or dammed and blocked off with carbon or graphite plates and carbon paste to seal any irregularities. The carbon plates and paste may, in turn, be held in place with braces or straps tack welded with nickel alloy electrodes. However accomplished, a complete edge seal under and around hot arc welds is absolutely essential to avoid molten metal run-out.

An example of area damming is shown in Figs. 6 and 7. The casting shown here was dammed on the face and back sides of the flange and welded from the edge. The sloping surfaces presented something of a problem (but not too much).

Metal run-out from the ends can be prevented in several ways. One is to tack weld steel dams in place (see steel cover in Fig. 7). A second is to have someone hold a steel bar or carbon strip in place to chill and hold the end metal of each layer in place. A third is to deposit a low amperage "bead" across the ends of each layer before depositing the hot fill part. Any of these methods will result in a rather rough and irregular surface that can readily be melted and washed off after finishing the fill part.

Control of Undercut

The finish of "Hot Arc" welds is crucial to the success of the process. There is a tendency for rather deep undercut around the edges of such welds and a tendency for the weld edges to become white iron and glass-hard unless the finish is properly handled.

The undercut tendency is due to the graphite flakes in gray iron and their not dissolving abruptly as the adjacent matrix metal melts under the intense heat of the arc. Thus, they tend to remain for a bit and for the intervening matrix metal to melt and draw away from the "dry," unfused graphite surfaced edges, due to surface tension in the molten material.

This situation can be handled in several ways. One is to leave the undercut and fill it with nickel alloy electrodes after completion and cooling of welds. A better method is to finish the edges by gas welding, which process applies heat at a slower rate and allows time for the graphite flakes to dissolve in the molten weld metal and for the edges to smooth out without undercut.

In most cases, other than with very coarse graphite, the undercut can also be eliminated by either carbon arc or gas tungsten arc welding the edges, as both processes permit slower melting and continued application of heat in a given region. If suitably experienced men are available, however, the gas process is generally preferable for edge finishing.

Controlled Cooling

Weld edge hardness is a function of cooling rate during the first three or four minutes after the deposition of passes or completion of welds. If weld edges cool from molten or near white heat to a dull, dark red or black heat within a minute or so, the edges are apt to become white iron and very hard. On the other hand, if, at the finish of welds, the cooling rate of the weld and especially the edges is controlled by application of gas torch heat so as to remain quite red, say around 1600 to 1700 F (871 to 927 C) for a period of about 2½ to 3 minutes and the cooling rate of the weld region is then controlled by torch heat so that the weld region does not cool to a black heat—say, 1100 F (593 C)—in less than another 3 to 4 minutes, the weld and heat-affected zones will be soft, gray and readily machinable.

The latter two details of undercut fill and controlled cooling of weld areas are vitally important to the success and machinability of "hot arc" cast iron welds.

Preheat

Comment about overall preheat and cooling after welding is pertinent and follows. In many cases, overall preheat to the aforenoted 1000 F (538 C) temperature range is not essential. It was not needed for the part shown in the lead photograph, largely because of no internal cross-members. Were the part more complex, containing cross-members, more even or overall preheat would be essential. In many cases, a tapered preheat may be used, with full preheat in the weld vacinity, about half that at midpoint across the part and little or none on the

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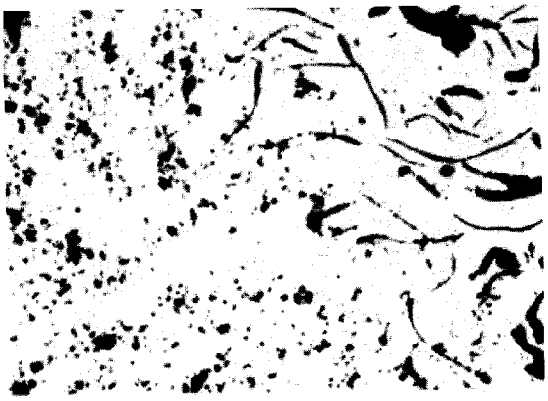


Fig. 10—Unetched "Hot Arc" weld and fusion zone structure. ×100

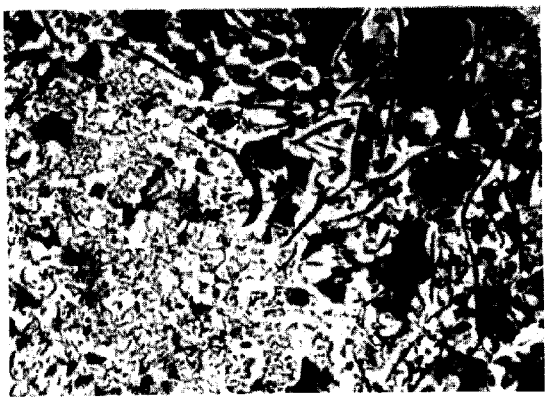


Fig. 11—Etched "Hot Arc" weld, fusion line and heat-affected zone structure. Weld is at left; base metal is at right. Nital etch, ×100

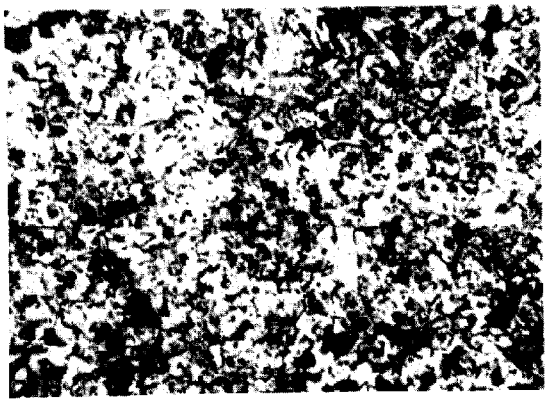


Fig. 12—Typical etched structure of "Hot Arc" weld deposits. Nital etch, ×100

opposite side. The same principle but carried to a lesser degree is usually adequate for most gray iron weld repairs—say, with a full 1000 F (538 C) in the general weld area—perhaps half that on the opposite side and gradually tapered heat in between.

Preheat temperatures alone without illustration of gradients may be somewhat misleading. Cast iron melts at around 2200 to 2300 F (1204 to 1260 C) which is a rather light red or cream colored heat. In welding, the fusion line must of course reach that temperature and there must be some sort of temperature gradient leading away therefrom. For soft, gray deposits, the gradient should not be too great.

Typical gradients for good deposit conditions are illustrated on Fig. 8. Although overall preheat may be either uniform and in the 900–1000 F (482–538 C) temperature range or tapered, immediately adjacent to weld areas, the base metal should be torch heated to a good red heat for the start of "Hot Arc" weld deposition—say, to about 1500 F (816 C)—with the drop to black heat of about 1000 F (538 C) spread over a distance of from about 8 to 12 in. as indicated by Fig. 8.

For maximum certainty of avoiding cracking or deleterious stress build-up, a full overall preheat followed by a furnace stress relief heat treatment at around 1000 F (538 C) is undoubtedly best. The large casting shown in Fig. 9 was handled in that way. It had over 200 lb of weld metal deposited in it in around 3½ hours one day, which illustrates the speed of the process. The casting was then returned to the furnace for a full overall 950–1000 F (510–538 C) stress relief. When machined, there was no indication or evidence of repair, which illustrates the quality of properly made "Hot Arc" cast iron repairs.

Discussion

The "Hot Arc" process for welding cast iron is not new. It was briefly outlined in the 1938 edition of the AWS Welding Handbook. It has also been referenced in Russian and other foreign literature. However, these references have been mainly in connection with electrode coatings to promote graphitization or produce gray iron deposits from electrodes with steel cores, details and references for which I no longer have.

Covered electrodes for this general type of cast iron welding were also available for many years in this country from the Chicago Hardware Foundry Company which, however, no longer makes the electrodes. The guidelines for welding with the Chicago Hardware Foundry Co. covered electrodes were, however, somewhat different from "Hot Arc" procedure. Lower amperage, smaller electrodes and lower preheat were generally recommended, along with more conventional quick solidifying bead deposition. From trial and what I have seen, such bead deposition frequently results in some sort of weld deposit trouble—either from slag entrapment, hard areas in welds, porosity, cracking or other factors. Conversely, high amperage, big puddle, hot weld region welding and controlled cooling procedures with "Hot Arc" have given consistently satisfactory results. So deposited, welds which are completely free of cracks and of full base metal strength can readily be made.

The metallurgical structures typical of "Hot Arc" weld deposits and heat-affected zones in gray cast iron are shown in Figs. 10-12. Tensile strength of such joints made with ordinary bare cast iron gas welding rods such as are available from any local welding supply house typically run from about 30000 to 35000 psi (241.3 MPa), and higher strengths should be readily attainable with appropriate alloying.

Conclusion

It is to be emphasized that the "Hot Arc" process is not intended for miscellaneous, odd-shaped or small part welding. It can only be used in the flat position with complete edge and root seal. For appropriate work, however, it offers an outstandingly fast method for making top quality, full base metal strength repairs in gray iron castings.

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GRAY CAST IRON WELDING

CALLS FOR APPROPRIATE PREHEATING MEASURES BEFORE PROCEEDING TO EFFECT REPAIRS WITH EITHER OXYACETYLENE OR ARC WELDING



Author Klimek using oxyacetylene torch and rod to make welding repair in gray iron casting.

BY J. KLIMEK AND A. V MORRISON

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The Iron Casting Society—located in Cleveland, Ohio—has provided data indicating that, in 1975, 10,622,000 tons of gray iron casting were produced. The production of ductile iron castings reached 1,825,000 tons, while 730,000 tons of malleable iron castings were produced. Data for white iron and alloy iron castings were not available.

We can reasonably assume that 0.1% of the gray iron castings produced either in 1975 or in previous years will fracture or wear in critical areas and can be salvaged by welding. On this basis, the task of weld repairing—its magnitude—and the economic advantage of reclaiming 10,662 tons of gray iron castings becomes apparent.

Cast Irons—Types and Properties

Generically, the term "cast iron" encompasses a broad variety of engineering materials, all of which have unique properties for specific applications. A brief description of cast irons—their types, physical properties and engineering applications—is pertinent to an understanding of the welding techniques used in their successful repair. In this respect, cast irons are differentiated by their color when fractured into two broad classes—gray and white. Also, they are further differentiated by microstructure, physical properties and alloy composition into many other grades. In turn, the physical properties of cast irons are established during the melting cycle by control of composition. If the iron is not an alloy iron, control is primarily related to the percentages of carbon and silicon in the alloy.

Gray Iron

The fracture surface of gray iron is gray in color due to the presence of free graphite. The alloy contains varying amounts of carbon, silicon, manganese, phosphorus, sulfur, and residuals and may be alloyed with other elements in small amounts such as nickel, molybdenum and chromium to enhance specific properties such as machinability, tensile strength, heat resistance, etc.

In addition to color of fracture, gray irons are further differentiated by specified tensile strength, deflection and brinell hardness, there being essentially no practical yield strength, elongation or reduction of area.

The rate of cooling between heavy and light sections and other factors in the mold can, for practical purposes, affect the physical properties of a gray iron casting. All of these variables must be controlled by the foundry to produce a satisfactory product.

White Iron

The fracture surface of white iron is white in color due to the absence of free graphite, all of the carbon being combined with iron or other carbide forming elements. The basic elements in white iron are the same as in gray iron. The difference is that white iron contains appreciably less carbon and silicon, and if alloyed the amount of alloying elements added is much more than in gray iron to promote maximum hardness and erosion resistance.

White irons may be white throughout the fracture or white on the outer area and gray on the inner area depending on chemistry, section size and cooling rate. The white irons are generally considered to be **nonweldable** due to high hardness, high residual stress, low deflection plus engineering applications and economics that dictate replacement rather than repair.

Malleable Iron

Malleable iron is a low carbon, low-silicon cast iron. It has definite physical properties of tensile and yield strength and elongation due to either a ferritic or pearlitic matrix with free carbon in the temper form. The temper form of carbon has the appearance of an irregular shaped mass; this gives the malleable iron properties closer to nodular carbon iron than to flake carbon iron. The microstructure and temper carbon form are developed by heat treatment.

Malleable iron is **not normally welded** although it can be brazed or silver soldered.

J. KLIMEK is Owner and Welder, Weldize Corporation, Brooklyn, N. Y., and A. V. MORRISON is Chief Equipment Inspector, Chevron Oil Company, Perth Amboy, N. J. (Editor's Note: At age 75, Mr. Klimek is still welding and has 50 years of gray iron welding experience.)



Fig. 1—Preheating a gray iron casting with charcoal before proceeding with welding repair

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Alloy Cast Iron

The term "alloy cast iron" can be construed to include a great variety of cast irons, from ductile cast iron to austenitic cast irons, including alloyed gray and white cast irons. The most unique of these is ductile cast iron which is made by inoculating gray cast iron with magnesium. The addition of magnesium causes the normal gray cast iron flake graphite to assume a nodular or spherical shape. Composition control and heat treatment result in a cast iron having the essential physical properties of a low carbon steel. The nodular structure is **not retained if** these irons are **welded.**

Other alloy cast irons encompass a range of alloy content from less than 1% to over 20% of alloying elements such as tellurium, phosphorus, nickel, chromium, molybdenum and silicon, each of which implants unique properties of hardness, heat resistance, corrosion resistance, and impact strength, etc., for specific engineering applications. Some alloy irons can be welded but normally are not.

Welding Repair of Gray Iron Castings

"Burning In"

If welding can be construed as being "the application of sufficient heat to a base metal to cause it to melt and form an alloy with an applied liquid metal or itself", then the first gray iron welding repair method would probably be the foundry method of "burning in" a defective casting.

"Burning in" is done by building a containment around the defective area and then slowly but continuously pouring molten metal into and overflowing the containment until, by judgement and experience, fusion with the base metal is achieved.

This type of weld repair is a foundry method applicable to relatively large expensive raw castings with small defects. It is not applicable to field or shop repair of finished and engineering applied castings.

Arc and Oxyacetylene Welding—Comparisons

Gray iron castings may be repaired in the field or shop by means of arc or oxyacetylene welding. However, either type of welding without proper preheat involves the strong possibility of cracking due to the concentration of heat resulting in severe thermal gradients and stress concentrations during the heating and cooling cycles. In addition, some section of the fusion or heat-affected zone matrix will transform to martensite which is an embrittled phase, thus compounding both current and future woes.

Of the two methods of welding without specific preheat, arc welding is more prone to cracking and martensite formation than is oxyacetylene. In addition, arc welding does not give a color match due to core wire, an iron-nickel alloy, being a different composition than gray iron, an iron, carbon, silicon alloy. This may or may not be of importance depending on application. The same is true of brazing with copper base alloys.

With either oxyacetylene or arc welding, preparation of a defect area by mechanical means—grinding, chipping, machining—is a prerequisite to success. Moreover, with arc welding, skip welding, peening and cooling to ambient temperature before applying the second pass, etc., are recommended to help avoid the undesirable effects of cracking and martensite formation due to the

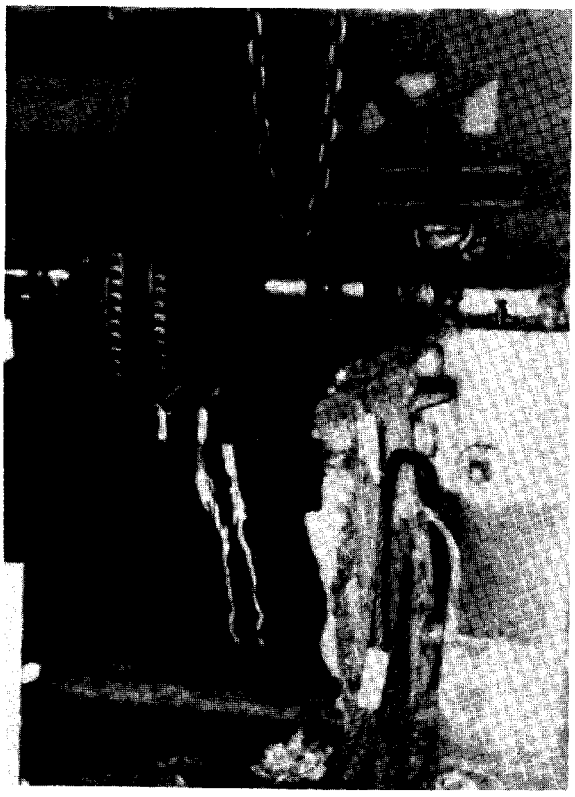


Fig. 2—Cracked "Elliott" 600# steam turbine



Fig. 3—Cracked "Elliott" 600# steam turbine prepared for welding repair and preheating

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heat generated by the arc. An oxyacetylene flame is considerably colder than an arc, and its heat is not as concentrated. The result is that a welder must apply considerable preheat before he can commence actual welding, a result being that cracking and martensite formation may be minimal.

Preheat Procedures

Some Basic Considerations

Preheating a gray iron casting sounds simple but actually requires forethought, equipment, time, patience and—very often—experience. Forethought is required to look at the defective casting and form a mental or written plan of job progress. Included in the plan are items such as casting alloy composition, overall dimensions of the casting, wall thicknesses of both the area to be welded and other areas, and grinding out the defect. Other items to be considered are rate of heating, positioning of casting for desired sequence of welding (which may involve several moves of a hot casting or cycles of heating-cooling-heating), electrode and torch sizes and types, fitting, thickness and type of insert material, etc.

The minimum preheat temperature to successfully weld gray iron castings is 700 F (371 C); 800 F (427 C) is preferred and up to 1100–1200 F (593–649 C) will not hurt and may actually aid in acetylene welding because some casting thicknesses are tremendous heat sinks, and proper fusion as well as impurity and gas bubble flotation require that the welder be able to maintain a pool of fluid metal.



Fig. 4—Welded "Elliott" 600# steam turbine

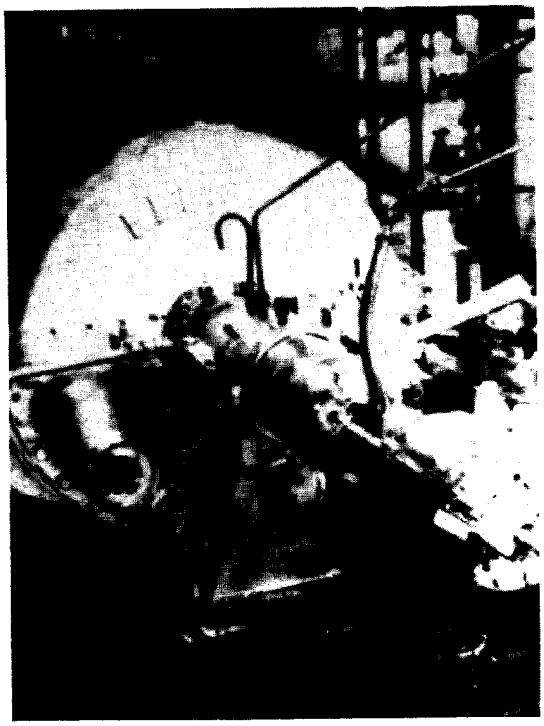


Fig. 5—"Elliott" 600# steam turbine back in service—only minor machining was required after welding repair

Equipment

Preheat equipment varies from the simple and cheap to the elaborate and expensive. All preheating should be done in a refractory oven. The oven can be constructed of loose bricks to accommodate one-of-a-kind type jobs, or it can be a permanent oven to accommodate multiple, similar jobs such as cylinder heads.

The use of a refractory oven is crucial to success when weld repairing complex castings. This is because it permits controlled rates of heating and cooling to avoid thermal stresses and retains residual heat during the actual welding operation.

Depending on the job, other equipment would include welding supplies, cranes, machine tools and miscellaneous handtools.

Fuels

The fuel for preheating may be gas, charcoal briquettes or wood charcoal. If the fuel is gas, it must be diffused through a bed of refractory briquettes to permit slow, even heating. With all but the most simple uniform-wall thickness castings, a direct gas flame will result in high thermal stresses and potential cracking.

Many old time practitioners of the art of acetylene welding gray iron castings (including coauthor Joe Klimek) firmly believe that the best fuel for miscellaneous small to medium sized castings is wood charcoal. However, a more modern concept well established in the pressure vessel, piping, chemical plant and petroleum refining industries which has not been applied to gray iron welding is to use electricity for fuel, especially when preheating large castings.

The development of reliable, accurate thermocouples and resistance heating blankets applicable to almost any

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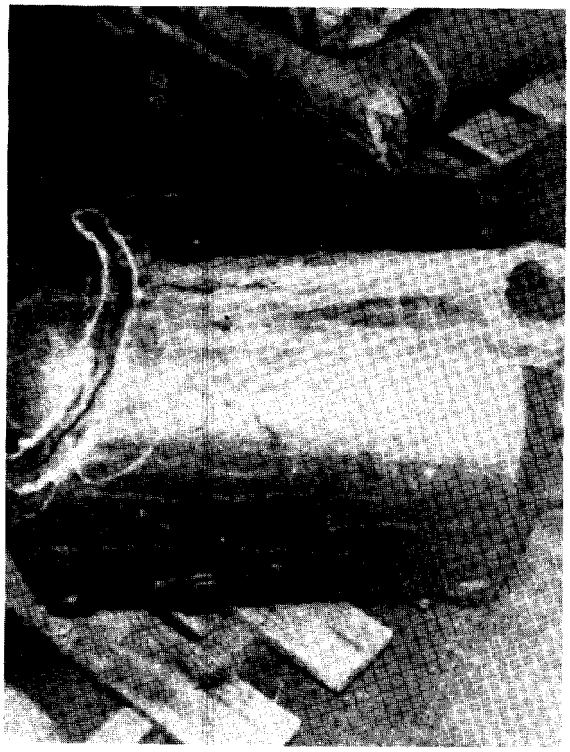


Fig. 6—Cracked "knock out pot"

shape, together with programmed multi-point time-temperature controllers and lightweight insulation, must be noted. Together, these are more feasible, safer and more desirable than an open flame for preheating prior to effecting weld repairs in gray iron castings.

Final Considerations

Preheating is time-consuming and is thus an anathema to much modern industrial production thinking. A multi-wall complex casting of varying wall thicknesses may take two to four hours to preheat slowly to welding temperature. To go much faster is to invite trouble from thermal stresses. It is common practice to preheat on the night shift and weld on the day shift.

If a casting is properly preheated, then either oxyacetylene or arc welding, or brazing, will make a useable joint depending on application. If the preheating is not correct, none of the welding methods will be totally satisfactory.

The choice of welding method is dictated by job variables, i.e.:

1. Is color match important? If "yes," use oxyacetylene welding and gray iron rod. If "no," use either oxyacetylene or arc welding with nickel alloy rod.
2. Is out of downhand welding required? If "yes," use arc welding, although oxyacetylene welding with gray iron rod can be used very well vertically.
3. Are increased strength and ductility required at a high stress crack location? If so, use arc welding and nickel alloy rod.
4. Speed of welding—this is important on any job for cost control. However, actual welding time is generally self-administrative, because the nature of gray iron welding is such that once you start to weld you must finish. Other factors such as preparation, preheating, finishing and overhead are cost controlling. However, it is to be noted that, with an oxyacetylene torch with a big tip and two 3/16 or 1/4 in. (4.76 or 6.35mm) rods welded together to make one, it is possible to deposit filler metal very fast.
5. Are pressure tightness and uniform thermal expansion important such as on cylinder heads or steam turbine or pump cases? If "yes," the choice is oxyacetylene welding with gray cast iron rod. This is because the finished welded casting must be as uniform as possible.

Differences in chemical composition as related to elements that impart fluidity to molten weld metal (i.e., carbon, silicon, manganese and nickel) make the difference between "good" iron and "dry" iron, and strongly influence the ease of fusion. Contrary to a popular welder's belief, the previous service of the casting does not "burn" these elements out of the metal unless it is a high temperature application such as the surface of a steel ingot mold.

The normal difference in composition between the skin of a casting and the interior also may present difficulties in tying in the surface pass. However, the main difficulty is that some castings have included gas pockets, sand, gas producing inclusions, etc., which result in pressure test rejections unless carefully removed during the actual welding operation. In this respect, oxyacetylene is superior to arc welding because the welder has a chance to work out the defects. The oxyacetylene welding is thus enhanced by the skills of an experienced welder of high integrity and devotion to the art such that its superiority cannot be surpassed as a repair method for gray cast iron.

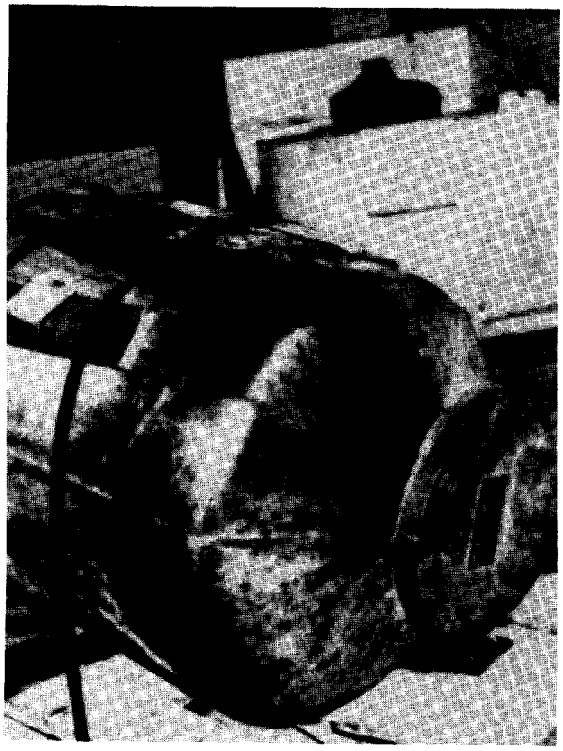


Fig. 7—Welded "knock out pot." No leaks were detected after welding repairs were made

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The metal arc welding of cast iron for maintenance and repair welding

by Clive Cookson, M.Weld.I.

The author argues that the development of low heat input weld metal compositions and welding techniques have made metal arc welding a viable proposition for cast iron repair work.

The metal arc welding of cast iron is still widely considered as a dubious method for joining and overlaying. It is true that many difficulties do exist, but with the development of specialised low heat input weld metal chemistries and welding techniques, these difficulties have been largely overcome, with the result that repairs over a wide field can be carried out confidently.

In this paper, the selection of suitable weld metal compositions is considered, and the paramount welding and metallurgical problems discussed, together with possible solutions. Finally, a selection of difficult but successful weld repairs is presented.

Heat input

The heat from the welding arc results in the formation of a liquid pool consisting of a weld/base metal mixture which brings a narrow band of base metal to an austenitic condition causing rapid solution of the graphite flakes or nodules.

This austenitic region becomes enriched in carbon, and the rapid quench effect from the base metal mass, as the arc is traversed, transforms the carbon-rich austenite to martensitic structures with eutectic carbides at the interface where carbon-rich microscopic pools have formed.

This interfacial carbide band can be continuous or discontinuous. The heat treatment necessary to transform this heat affected zone (HAZ) and so alter the structure is generally impractical on large components, and also on smaller components which have been extensively machined, due to scaling and distortion problems.

The depth of the HAZ is relative to the heat input and the welding technique used. It has been shown that where electrodes are used which exhibit a soft arc-brazing characteristic, minimal hardening occurs due to a reduction of hard transformation products.[1] This is illustrated in Fig. 1 with a macrograph of two weld deposits made under identical welding parameters. The low dilution value can be seen on the left, which has produced minimal hard microconstituents in the photomicrograph beneath. There is evidence of only slight, discontinuous carbide at the interface. In Fig. 1, right, high dilution has occurred with mass martensitic and carbide structures shown in the relevant micrograph.

Mr Cookson is Chief Welding Engineer at Eutectic Co Ltd, Feltham, Middlesex.

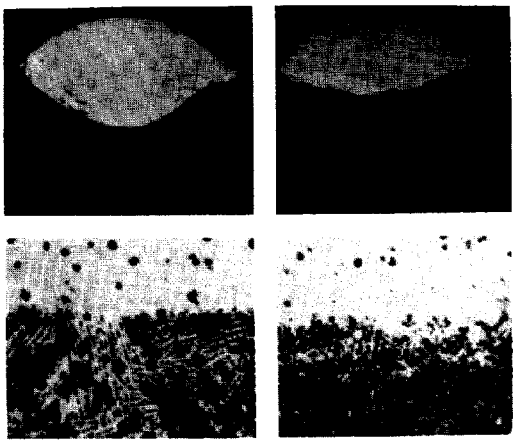
The depth of hardening of a particular cast iron, which governs the width of the HAZ, depends to some extent on the structure of the cast iron before welding. The presence of pearlite, especially in SG irons, can influence the depth of the HAZ, the controlling factors being solution of carbide and diffusion rates of carbon. The actual hardness of the HAZ, however, does not differ very much with different irons.

Choice of electrode chemistry

Research work is being carried out by the author's company on improvements to metal arc electrodes for the welding of cast iron. It is generally established that the 60/40 nickel iron type of electrode offers the highest crack-resistivity, but remarkable results have been obtained with various flux coatings extruded on identical filler wires. Flux densities, remelt and flotation characteristics are also pertinent factors in the homogeneity of the weld metal.

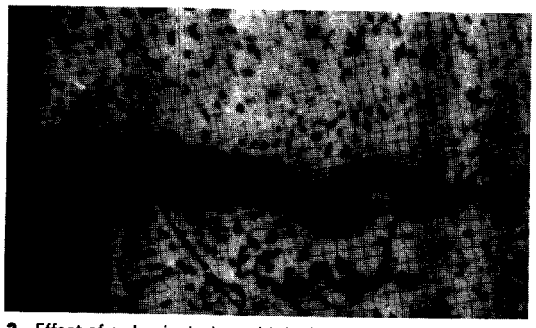
Figure 2 shows the effect of a slag inclusion and the incipient hot cracks emanating from this. Microscopic inclusions of this nature can lower the tensile strength to around 93 N/mm^2, as shown by specially prepared tensile specimens.

1 Two weld deposits made under identical welding parameters: (left) low dilution value, (right) high dilution

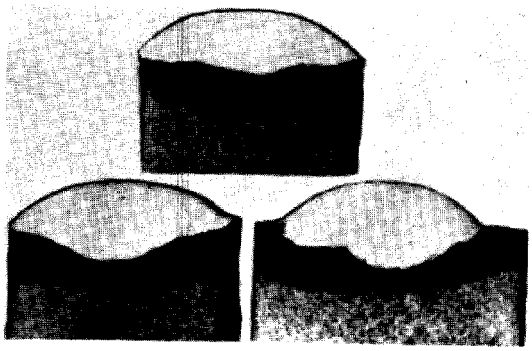


Source: Metal Construction and British Welding Journal, May 1971 (The Welding Institute)

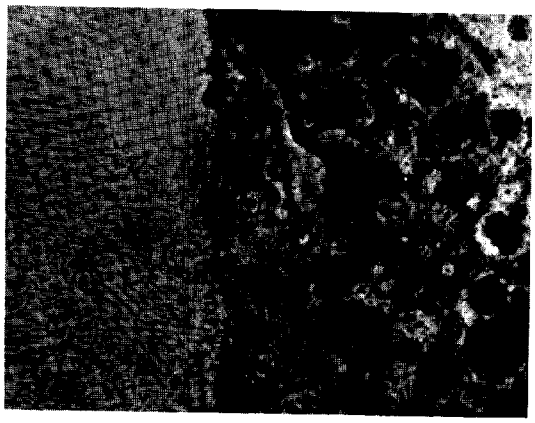
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2 Effect of a slag inclusion with incipient hot cracks (× 100)

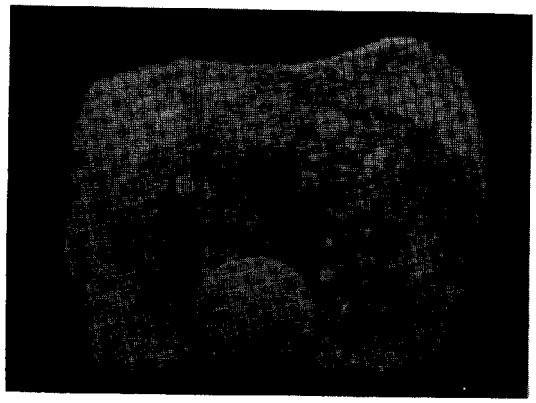


3 Effect of three different coatings on a 60/40 Ni:Fe alloy using identical welding parameters



4 HAZ cracking, using austenitic stainless steel filler (× 100)

5 Formation of massive brittle Al:C:Fe nodules, aluminium bronze (× 10)



The test programme indicates that an even wider range of electrode types with repetitive core wire chemistries will be required in the maintenance field of weld repairs to cope with specific problems. This could well illustrate the folly of grouping electrodes by core wire composition, and making a selection on economic grounds. (Figure 3 shows the effect of three different coatings on a 60/40 Ni:Fe alloy using identical welding parameters.)

Austenitic stainless steels

An austenitic stainless steel is sometimes favoured for the welding of cast iron where bonding problems are encountered. Although only a low percentage of presumably successful repairs have been made using this chemistry, it usually involves the deposition of only a small amount of weld metal. However, even a few successes will raise an obvious danger if this opinion becomes more widespread.

Investigations by the author have shown that in certain cases successful welds can be made on cast iron, using austenitic stainless steel electrodes, where the degree of restraint is low, or where the deposit is in compression to the base metal.

AWS E312/16 filler metal, popularly referred to as the 29/9 type (Cr:Ni), was also employed in the test programme; this seemed an obvious choice due to its high crack resistivity attributable to the high proportion of intergranular δ-ferrite which provides preferential precipitation sites for the dissolved impurities and certain elements melted from the base metal.

Various degrees of cracking were observed in the weld metal, due possibly to the excess precipitation of carbon at the ferrite/austenite phase boundaries. HAZ cracking was very prominent where restraint was inherent in the joint (Fig. 4).

HAZ cracking does epitomise the problem of using austenitic stainless steel filler materials; their undesirably high physical properties make the brittle HAZ even more pronounced and vulnerable to stress. The yield point of austenitic stainless steels *exceeds the strength of the lower strength cast irons.*

A further risk with the use of these welding materials is the introduction of chromium which increases the hardness of the austenitic transformation products and of the carbides formed, $Fe_3C \rightarrow Fe/Cr_3C$.

Tin bronzes were also investigated and three problems were confirmed: (a) weld metal cracking, (b) poor bonding, and (c) base metal cracking due to copper penetration of the grain boundaries.

Aluminium bronzes proved to be successful in many cases, especially with cladding operations on cast iron, but iron and carbon pick-up also proved to be highly detrimental on restrained joints, especially where higher than usual amperages were used on initial runs to overcome bonding problems, due to the formation of massive brittle Al:C:Fe nodules (Fig. 5). With heavy build-ups, cleavage of the initial run occurred, due to embrittlement in this area.

Tests using a nickel buffer layer and a low penetration technique (Fig. 6) minimised this problem. However, the high tensile strengths and low elongation values (< 5%) developed with the heat-treatable aluminium bronzes as-deposited, make them a dubious choice for the general repair welding of cast iron.

Ferritic deposits

The use of ferritic deposits for the welding of cast irons is frought with many problems. Figure 7 shows a grey

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cast iron joint edge clad with two layers of a ferritic deposit; the extent of cracking can be seen.

With single deposits, cracking was in many cases not observed, but the structure consisted of a broken carbide network in a martensitic matrix. The average hardness of the single layer was 500 DPN.

There are cases in maintenance repair welding where the choice of this chemistry is a viable one, but it is incorporated sparingly and in individual stringer beads, usually in chamfered grooves to obtain high strength bonding.

Nickel-base electrodes

Electrodes containing >90% Ni and the 60/40 Ni:Fe type have undoubtedly emerged as the most satisfactory chemistries for the welding of cast irons, generally because of their desirable mechanical properties and ability to precipitate the carbon picked up from the base metal in its free form as graphite.

Figure 8 shows a 60/40 Ni:Fe deposit, consisting of a cored, solid solution structure containing graphite nodules and aggregates, the nodular form due to residual magnesium in nickel, used as a deoxidant.

Certain weaknesses can exist in a deposit containing a microstructure in which free graphite aggregates predominate. A preference for the formation of graphite spheroids + carbide is indicated and this can be encouraged with modification to the flux coating.

The high nickel electrode, while being more ductile, is prone to weld metal cracking under restraint during welding. However, on large fills, the higher strength nickel iron deposits can create a tensile stress on the HAZ and promote cracking problems in this area. The author devised a 'cross' technique which overcame these problems on a specific repair, and this is described later.

Tensile tests carried out on SG cast irons welded with a 60/40 Ni:Fe electrode, showed average values of 550 N/mm² on pearlitic structures, and 454 N/mm² on ferritic structures, elongation (50 mm) being 2% and 6% respectively.

In-service conditions

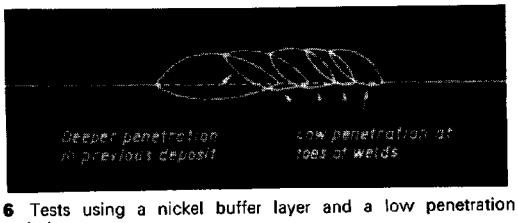
Grey flake cast iron has relatively low ductility and so its uses include in-service conditions which do not involve high fluctuating stresses or shock loading in which the SG irons are better employed. The welding of the SG irons for this kind of in-service condition is, however, not under discussion. Where tensile stresses are involved with grey iron casting then the framework is inevitably massive to provide a high safety factor.

Grey cast iron is widely used for machine tool castings, pump casings, bearings, heavy rolls, turbine casings, liquid containers and many more. There is a strong need in industry today to repair worn and broken parts to prevent downtime which is sometimes disastrously extended with long delivery dates. Without any doubt, the above applications constitute the largest proportion of cast iron welding.

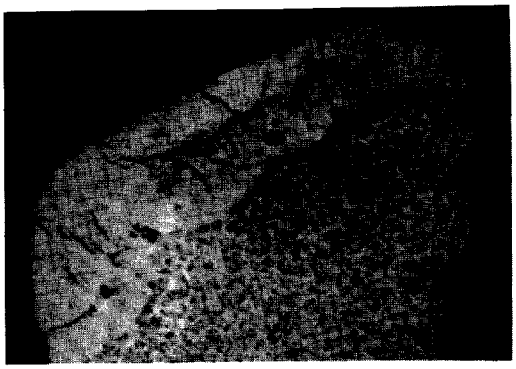
The problem facing the welding engineer is that the component should be restored to a mechanical condition which is comparable to the original casting. Many failures occur due to lack of knowledge of the problems which exist in the welding of cast iron.

The choice of weld metal chemistries has been shown to be vital, with the nickel-free chemistries providing minimum chance of success.

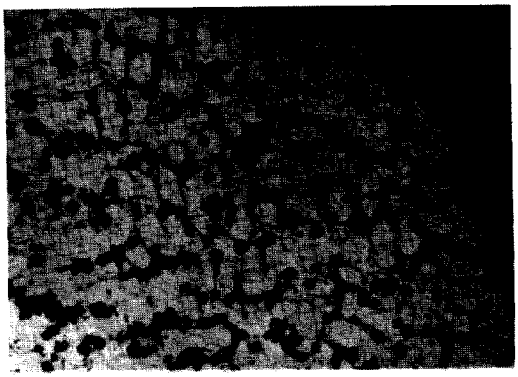
Failures which could have been avoided only strengthen opinion among some factions that cast iron cannot be welded with the metal arc process, and this could be no further from the truth.



6 Tests using a nickel buffer layer and a low penetration technique

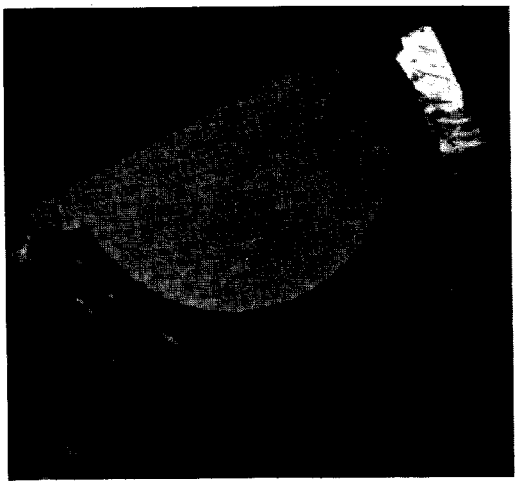


7 Grey cast iron joint edge clad with two layers of a ferritic deposit



8 60/40 Ni:Fe deposit consisting of a cored, solid solution structure containing graphite nodules and aggregates

9 Extent of wear in a segment of a steel half casing



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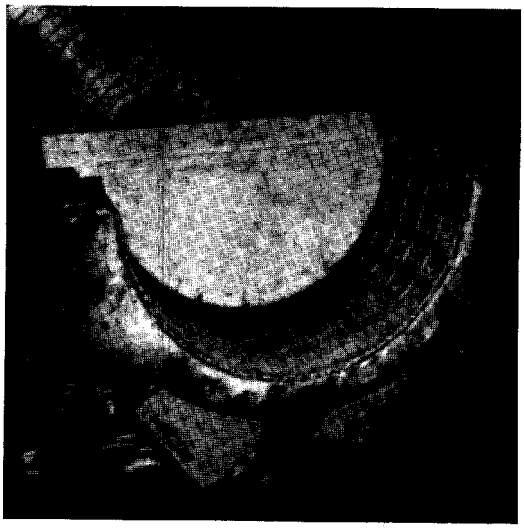
Problems in welding

The main contributory factors to failure of a weld repair to a casting during welding, after welding, or while the component is in service are as follows:

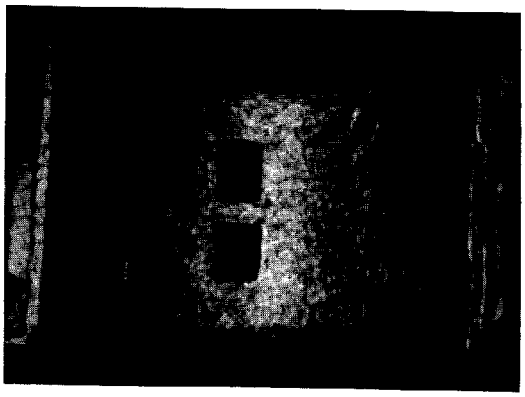
Ferritic deposits. Ferritic deposits will harden when deposited on cast iron due to carbon pick-up and the high quench effect. Apart from their sensitivity to cracking, they also develop undesirably high physical properties, including high yield strengths.

Heat build-up. There is usually a heat build-up in the casting unless long inter-weld pauses are observed. Expansion and contraction is unbalanced and cannot be tolerated on highly restrained areas of castings and especially thin sections. Grey cast iron has negligible ductility, and so the stresses developed cannot be absorbed by distortion in the casting.

The MS Mf range for cast irons depends on a number of factors, but the Mf point generally is not below 90°C. Tests have shown that if the weld deposit length is curtailed to × 2 the base metal thickness on thin sections, the temperature upon breaking the arc is below 100°C, falling in seconds to around 40°C on the first run. If an airline is employed to quench the deposit and remove the excess 20°C, then each and every run will be a facsimile of the preceding one. No detriment will result to the base metal by quenching from up to 90°C when this temperature is induced locally in the weld area with curtailed deposit length on a previously cold casting. The quenching of small castings would not be necessary where a uniform heat build-up can be obtained.



10 Technique of depositing stringer bead 'squares'



11 Half casing completed, showing extent of build-up

12 Stack cooler repair; a cross design was built in the centre of the defect



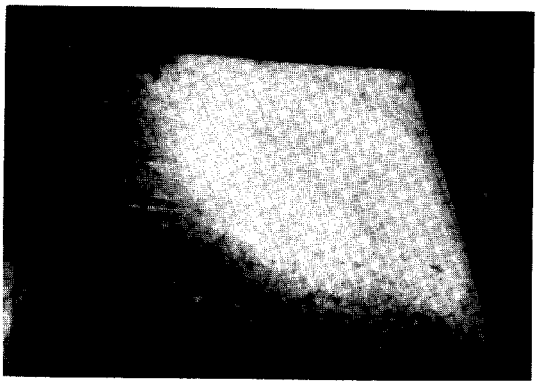
Crack susceptible weld metal chemistries. Deposits which have significant ductility dips and/or low melting intergranular films are not generally suitable for the welding of cast iron where restraint in the joint or repair area is encountered, this being frequently the case on massive castings.

Improper welding techniques. It is generally advisable to deposit stringer beads when welding cast iron, so avoiding an excessive weaving motion. Craters must be filled both at the start and completion of a weld run. A parallel line of stops and starts should also be avoided. De-slagging must be judiciously exercised to avoid flux entrapment.

Heavy build-ups. Large deposits create an undesirable tensile contractional pull against the base metal, and so it is necessary to drastically reduce this by peening and other techniques like the 'cross' formation with a more ductile chemistry.

HAZ hardness and width. Ways of minimising the extent and hardness of the HAZ have been discussed, as this is the prime reason for cracking and failure of welds on cast iron. It has been proved conclusively that cutting stress grooves and so scolloping the joint surface, and consequently the HAZ, together with the use of low heat input chemistries and techniques, virtually resolves this problem. However, the geometry, design and spacing of the grooves is of the utmost importance and can vary with each repair.

13 Stack cooler repair; line at weld/base metal interface caused by austenitic type filler metal and ferromagnetic casting creating north and south pole



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Preparation. The importance of correct preparation of the defect or damaged area before welding cannot be over-emphasised. The area should be ground and dye-penetrant tested to ensure a sound base, with all corners and edges radiused. The casting should also be warmed to 20°C before the welding operation begins, as hydrogen induced HAZ cracking in the welding of cast iron cannot be ignored. It is also advisable to bake electrodes immediately prior to use. On no account should damp electrodes be used.

Some successful weld repairs

Half casing. In one example, taken from a major steelworks, extensive wear had resulted in severe damage to four bearing half casings from a large sheet steel cropping machine. The material was grey flake cast iron to BS 1452 Grade 14. Figure 9 shows the extent of wear on one of the segments. Numerous welding contractors were approached and all considered the repairs to be impossible.

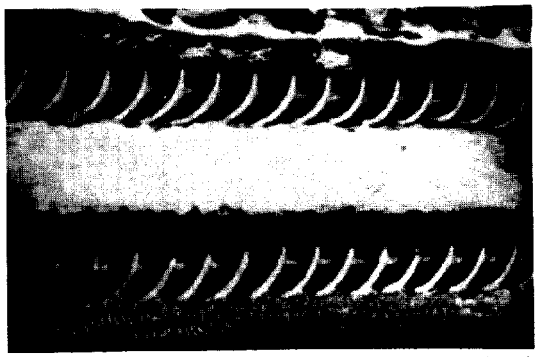
A procedure was drawn up by the author which included preparing the surface and cutting stress grooves across the face. The grooves were filled with a low heat input electrode (60/40 Ni:Fe type). The area was then divided into squares, using stringer bead deposits (Fig. 10). These squares were then filled with the deposits transverse to those in the adjacent square to break up the stress pattern. A high nickel type deposit was also employed in the repair, observing techniques previously discussed.

Figure 11 shows one half casing completed, and the extent of build-up can be seen. The nickel–iron type deposit was used on the final layers because of its better bearing properties. The repairs were completed eighteen months ago and the components are still performing satisfactorily.

Stack cooler. In another example, unavoidable sand scabs during the casting operation had resulted in defects in grey cast iron cooling panels for furnace lining in steel production. The section thickness of the panels was 125 mm, and the defects averaged 150 mm in diameter and up to 37 mm deep. It was important that the thermal transfer properties were not denuded by weld defects including cracks or porosity. Earlier attempts at a repair resulted in weld metal cracking and weld/base metal cleavage, the latter caused by an undesirable tensile contraction stress in the weld deposit.

The following technique was devised and proved completely successful in the nineteen repairs. The area was ground and dye-penetrant tested, and then warmed with a flame to 20°C. Stress grooves were cut around the periphery, and filled with a low heat input metal arc 60/40 Ni:Fe type deposit. The low penetration buttering technique was carried out, followed by grinding and testing. A 'cross' design, using a high nickel electrode, was built in the centre of the defect (Fig. 12) and the intercross areas filled with the 60/40 Ni:Fe type previously used. Grinding of the surface and dye penetrant testing was observed at each fill-in stage of this first test.

The 'cross' technique was used on each layer built up, ensuring that the crosses were in different positions to avoid a straight-through layer of high nickel deposit. The final layer was ground and tested both with dye penetrant and the magnetic particle technique. Note the line at the weld/base metal interface, Fig. 13, caused by the austenitic type filler metal and ferromagnetic casting creating a north and south pole in much the same way as a discontinuity.



15 Ship's turbine repair, showing extent of cavitation (above), pad layer ground to size (centre), and (below) descaled area ready for welding

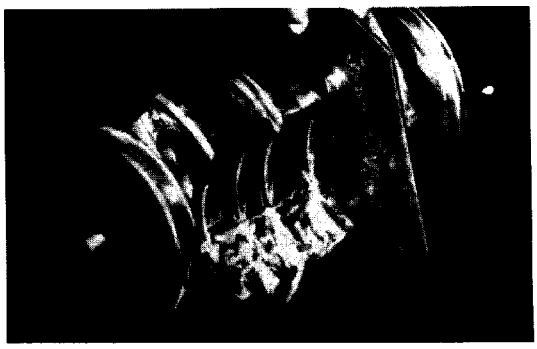


16 Stress grooves filled with a 60/40 Ni:Fe type electrode

14 Difficult working conditions for ship's turbine repair welding



17 Effect of peak technique for required build-up (peaks indicated by chalked arrows)



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18 Finished repair after filling in inter-peak areas and grinding to size

Ship's turbine. A few years ago a weld repair operation was successfully carried out to two LP cast iron marine turbine casings on board the 28,000 ton P & O liner RMS Oronsay. The repair work was necessary owing to extensive metal loss, caused by water droplet erosion resulting from high velocity impingement of moisture beads expelled from the rotor blade tips. The casings were grey flake iron conforming to BS 1452 Grade 14, and the blades were of Monel.

Preferential loss of the cast iron base metal had occurred (a) on the surface and inside edge of the gasket face, (b) between the individual blades on the first rows of each casing, and (c) the spaces between the blades up to and including the first eight rows. The root lacing wires were also exposed on the first rows by loss of the low moor iron locking strips and spacers.

Metal arc welding was employed, using a 60/40 Ni:Fe soft brazing type electrode to replace the lost metal, to consolidate the blades to the casing, and to fill in the inter-blade areas where loss of the spacers had occurred.

The quench weld technique, using an airline, was employed with peening of the deposits; 90% of the welding was carried out in the vertical-up position. The success of this repair resulted in a similar repair being carried out on the RMS Himalaya. The Oronsay repair has been described in detail elsewhere.[2]

Figure 14 shows the difficulties of the working conditions; Fig. 15 shows extent of cavitation (above) pad layer ground to size (centre), and (below) a descaled area. The need to obtain sound deposits can be appreciated as cleavage of the weld metal could cause extensive damage to the blades of the casings and rotors.

Wobblers in steel works. The wobbler ends of cast iron and steel rolls are subject to extensive wear. In this example, while welding of the steel wobblers poses no real problem, the SG cast iron wobblers had not been successfully reclaimed. The techniques outlined previously were conformed to, with air quenching exercised throughout the repair.

Stress grooves are cut across the worn face and then filled with a 60/40 Ni:Fe type electrode (Fig. 16). Due to the extensive build-up required, a 'peak' technique was devised (Fig. 17) which simulated small mountain peaks, indicated by arrows. The inter-peak areas were finally filled and the repair ground to size (Fig. 18). Four such welding operations are required on a single wobbler end. The repair gave eighteen months satisfactory service.

The author does not deny that the welding of cast iron is a difficult and complicated task. It requires specialised knowledge and, possibly, a welding technique tailored to each individual problem.

Welder training and absolute participation in the repair lay-out is vital. Nickel-free chemistries can, and indeed are, used in the welding of cast iron, but any preference for these should be calculated against a possible failure which would make the nickel-free types the most expensive ones.

Massive repairs on sections up to 1200 mm thick have not been discussed in this paper, although such repairs frequently have been carried out. One such repair to a massive cast iron frame in a steel works has been performing satisfactorily for over two years, with high tensile stresses on the joint area. However, techniques for these repairs are radically different from those that have been discussed, and fall outside the scope of this article.

References

1 Cookson, Clive. 'Repair of LP cast iron marine steam turbine casings'. *Weld. and Met. Fab.*, August 1969.
2 Cookson, Clive. 'Quench welding process for joining cast iron'. *Weld. and Met. Fab.*, August 1970.

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Maintenance and repair welding of castings

By C. Cookson, FWeldI, LIM, AMIET

Well over 85% of all weld repairs in the field of service repair and maintenance welding are carried out on castings. In many respects the repair and maintenance technological side has to be highly alert, because modern engineering developments constantly create gaps in the acquired knowledge, gaps that have to be closed with the aid of welding engineering and metallurgical work. New concepts of weld metal chemistries also have to be developed to meet these contingencies and to maintain the repair and maintenance field as a specialised operation, even though largely operating outside that line of welding controlled or guided by national or international standards. Full metallurgical backing is just as essential as welding engineering, and both these functions are usually controlled by one man, the Welding and Metallurgical Engineer. His job is to examine a damaged component, assess all the facts, carry out the required tests, then draw up a welding technique and choice of welding material and often do the practical repair himself. Quality control of all welding products brought into the company may be his responsibility also, together with the development of custom-made weld metal chemistries and fluxes. Unfortunately a repair technique used on one casting is not a panacea: each repair must be considered individually. The case histories discussed in this Paper merely brush the myriad of problem repairs in the repair and maintenance field.

CASE HISTORY NO. 1

Metal-arc repair to cast iron bridge

In 1968 serious cracks were observed in the structural members of the Osney Bridge, Fig.1, which carried the A.420 over the River Thames in the western approach to the City of Oxford. The structural beams are cast iron similar to BS 1452, Grade 14. The cracks had propagated to the fillet area from focal stress points and in some instances had encroached into the cruciform section of the upright. Seventeen cracks were found, all similar in magnitude, every one on the side nearer the crown of the bridge section.

The exact cause of the damage was not known, but it is assumed to be associated with cyclic loading caused by the passage of heavy vehicles during the war periods, and the

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presence of large surface sand scabs in an area adjacent to the focal stress points.

In view of the extent of cracking a repair was necessary, and, because of the condition of the parent metal, it was necessary to effect the repair by metal-arc welding. A high safety factor was achieved by incorporating into the weld repair plain carbon steel bars, which extended across the front face of the web, thus giving an additional bond to more sound areas.

A section of the parent metal was removed for analysis which showed the phosphorus content to be 0.9%. A special welding procedure and technique had to be devised to minimise weld metal repulsion and to eliminate lack of fusion and cracking dangers.

The bridge, which was completed in 1888, consisted of eight arch ribs, two sections of which were more recent additions to widen it; these were sound. The remaining seven inner arch ribs contained 112 'junctions', seventeen of which had suffered extensive cracking.

A two-stroke, 200A, 450 cycle, AC welding machine was positioned on the bank, and the earth lead was fixed to the outside section of the bridge. The welding lead, which measured 100m was laid across the river bed and brought up into a barge to a receiving plate from which two electrode holders were taken.

In view of the amount of inter-deposit work of peening, deslagging and cooling, it was possible to have uninterrupted welding from both leads, though not simultaneously.

All the members of the bridge were ground and dye penetrant tested, Fig.2a. A 6mm diameter hole was then drilled at the termination of each crack which was afterwards cut through with a hacksaw, Fig.2b. The cracks were ground out U fashion, one-third from each side, and a nickel strip inserted. The insertion of the nickel strip was a new technique devised by the author, Fig.2c.

The electrode used for the entire repair was a 60%Ni-40%Fe type, of 3.2mm diameter, as this allowed better practical control during welding which was carried out in the flat, horizontal/vertical, and vertical up positions.

It was necessary, when welding began at 0800hr, to use a propane torch to warm the parent metal to $>30^{o}C$ to avoid condensation and any danger of hydrogen-induced heat-affected zone (HAZ) cracking.

The technique of inserting the nickel-iron strip into the hacksaw cut through the crack may be summed up as:

1 The hacksaw cut removes any rust and impurities which may cause outgassing during welding, with subsequent disruption of the welding arc.

2 The metallurgical notch effect at the root is possibly minimised.

3 Practical tests and metallurgical examination of typical sections showed clearly that the weld metal on the initial weld beads on the outside of the joint was allowed to contract by the closing of the joint faces into the annealed nickel strip.

4 Although no significant additional strength would be achieved in the joint by welding each side into the nickel strip, the neatness of the joint and other attendant advantages can be appreciated.

The bar-reinforcing operation was preceded by a weld cladding operation. A 12mm diameter 0.4% carbon steel bar, bent to fit the configuration, was then positioned and fillet welded to the clad layer, Fig.3. The yield strength of the bar and weld metal was calculated at $>460N/mm^2$ which added a high safety factor to the repair.

It is suggested that in view of the excessive sand scabs, which had to be ground away at the joints, this repair was possible only by metal-arc welding. The success of this repair, after more than three years, may be of real significance, as this is, to the author's knowledge, the first of its kind.

CASE HISTORY NO. 2

Repair of a furnace roll

Attempts at weld reclamation on 25/12 Cr-Ni furnace rolls (C=0.5%) from a Priest normalising furnace, which had been in service for some years, were unsuccessful because of the severely embrittled state of the parent metal; this had resulted in extensive cracking in both weld metal and parent material. The preweld solution heat treatment was not possible with the equipment available. The author was asked to investigate the problem.

The roll, which measured 5.5m in length, 419mm in diameter, with a wall thickness of 38mm, was constructed from a barrel, centrifugally cast in one piece, with a bell-end casting (static cast) welded into each end after shrink-fitting.

During service, the maximum live load of 1783kg is distributed over a 3.2m length, and the maximum temperature was estimated at $960^{o}C$, although this ceiling temperature

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depended on the specific heat treatment being carried out. Each roll is driven from one end at 1rev/min.

Service failure of the rolls was caused by extensive cracking in the weld/parent area of the original fabrication, a situation which often results in complete separation of the bell-end casting from the barrel during service. This particular roll had been in service for seven years, and the problem seemed relevant to all the other rolls in the furnace as regular replacement was becoming necessary.

Preparation of weld area

Complete removal of the cracked and eroded weld metal and adjacent parent metal to provide a joint design free from defect, produced a channel 127mm wide; this promoted a condition which underlined the stress problem because of the extensive weld deposit then required.

A section of sound parent metal was removed from the joint area for metallographic examination. The structure, Fig.4, is basically austenitic with prolific intergranular M_{23} C_6 carbide, which has coarsened, presumably because of the relatively long periods at high temperatures. The presence of sigma is indicated by the random needle-like phase.

Further sections were taken of weld deposits, tests being carried out on joint faces in the repair area using conventional welding metal. The 'bead-on-plate' test was also conducted on various parts of the joint, and gross fissuring occurred around the entire periphery of the deposit.

The tests proved that a conventional weld repair was totally impractical, especially in view of the extensive joint preparation necessary to provide a sound parent area.

The embrittlement of high chromium and austenitic chrome/nickel steels by the formation of sigma and M_{23} C_6 carbide can occur when service temperatures in the range 680°-850°C are encountered. Although the degree of embrittlement may be extensive in many situations, experience indicates that the component can still continue to give satisfactory service, although any attempts at weld repairs are largely unsuccessful because of parent metal cracking caused by nonuniform expansion and contraction, and the stresses created by weld metal contraction.

Although the maximum operating temperature of the furnace, 960°C, is above the upper critical temperature range for the formation of sigma, it must be realised that such a furnace may frequently operate or be held for long periods within the 600°-850°C range, and once the sigma phase has been formed the maximum operating temperature of 960° is too low for rapid dissolution.

Problems to be overcome

It was established from weld deposit tests that a number of problems had to be surmounted for a successful repair to be obtained.

Firstly, parent metal embrittlement, presumably caused by the prolific formation of M_{26} C_6 carbide and some sigma, reduced the parent metal ductility, and the stresses created by localised expansion and contraction, because of heat from the welding operation, could not be absorbed without cracking occurring.

Secondly, the extensive joint preparation required demanded a prodigious amount of weld metal which would produce very large strains on the joint walls.

This particular problem of embrittled cast austenitic steels is synonymous, from a practical point of view, with the low ductility found in certain grades of grey cast iron, although to a lesser degree. Newly developed techniques have now largely overcome this problem in welding cast iron.[1] The technique used involves depositing short stringer beads of weld metal which are rapidly quenched with an air line with extensive peening of the weld beam immediately the arc is extinguished. The technique achieves:

(a) The peening operation which should reduce the height of the weld bead by one-third, using a three-pound ball-pane hammer, minimises the weld metal contraction stress which may be serious in large fills.
(b) Quenching the weld bead maintains the casting at room temperature, thus alleviating the serious problem of localised expansion and contraction.

Extensive bead-on-plate tests were then carried out, observing the air quench and peening techniques, and fully crack-free deposits were obtained. It was, therefore, decided to proceed with the repair.

Repair technique

To minimise the reaction stress the entire joint area was clad, using a 'soft arc' coating 25/20 Cr-Ni type metal-arc deposit. A 3.2mm diameter electrode was employed to limit the region of contraction strain, before peening of the deposit reversed this stress to a compressive one to the base metal.

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After the joint faces were completely clad a close inspection revealed no evidence, at this stage, of cracking in the weld or base metal. It was then anticipated that a limited degree of ductility would be afforded by the clad layer against contractional stresses which might exist when the joint itself was filled progressively.

The root deposit was completed next, using the back-step technique with air quenching and comprehensive peening. It was decided to incorporate 19mm round bar sections in the repair in a 'Jacob's Ladder' formation to reduce the amount of weld metal required by 6kg and thus the overall strain, Fig.5a.

AISI Type 304 bars were the only bar material available and although it is inferior to the furnace roll material, it was adopted because it would be well below the surface layers. The bars were raised and tacked at each end to allow complete root penetration from both sides, still using a 3.2mm diameter electrode. The bars were then dressed down with a hammer, having previously been bent to the required shape to produce a consistent root gap. The repair proceeded using the same technique of depositing short runs, immediately peening with a simultaneous quench, using an air line with a jet forced as high as could be tolerated.

The joint was filled in the next stage using a block sequence with 5mm diameter electrodes, and was finished in the inter-block areas after the entire circumference had been covered to reduce the degree of stress in the joint. A final capping layer was deposited without peening, Fig.5b.

It is appreciated that, in some quarters, peening is considered to be detrimental to the corrosion resistance of an austenitic stainless steel, and also that the susceptibility to the formation of the sigma phase may be enhanced by this operation. However, peening was a technique which was found to be essential in this repair. The capping weld deposits induced a certain amount of stress relaxation in the underlying beads (which had been work-hardened by peening), and thus some benefit would be obtained, even though the transformed area would extend through only that portion of the remelted bead and HAZ produced in that same bead.

This particular repair was in service for about three years when it was removed for another cause. Numerous rolls have been repaired using the same technique.

CASE HISTORY NO. 3

Repair of wet liner seal in ship's engine

When a ferromagnetic cast iron is welded without preheat hard constituents are formed in the HAZ; these include carbides, martensite, and bainite. Where stresses are induced as a result of contractional strains, unless minimised by special welding techniques, this embrittled area can create cracking problems. However, cold welding is widely exercised, and is a viable method of repair and reclamation.[1] It is also a proven method of welding white cast irons.[2] However, it is incorrect to assume that a preheat should not be used whenever possible. In fact it is highly desirable to preheat pearlitic SG cast irons to around 250°C to avoid cracking, and even grey flake irons benefit enormously from some degree of preheat.

The repair to a ship's wet liner seal was a typical example of a repair where preheating could not be used. However, owing to the importance of the operation it had to be shown quite clearly that a sound technique was to be used. The area to be welded, Fig.6, was tested to ascertain the extent of the damage. The broken-out section, Fig.7, was also damaged and this was prepared and preheated to 400°C, then heavily clad on the joint edges and outside faces, where two high tensile bars were to be positioned. A postweld heat treatment of 900°C was carried out to dissolve the carbides in the HAZ.

The object of welding the broken section in this way was to achieve a mechanical joint in addition to the joint proper; Fig.8 shows the general idea. The joint faces of the main casting were then clad, using short runs and peening, and maintaining a localised interpass temperature of 100°C with large soft flame burners.

The patch was fitted into position and welded in the same manner, i.e. short runs and peening.

The two bars (12.7mm diameter) were of En 110. These were initially clad with the TIG process and a nickel wire containing titanium (employing a 250°-300°C preheat) to allow the relatively cold welding of the bars to the casting at a later time, and so avoid HAZ cracking in the bars. Figure 9a shows the completed repair with one bar visible.

The final stage of the repair was the tinning of the inside face of the broken section, using a tin-based solder having a high melting point. This was then scraped in to within 0.025mm using an accurately ground template, Fig.9b.

It will be appreciated that, apart from a sound weld repair, a mechanical joint of some strength was achieved, and in addition there

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were two high tensile bars overlapping the joints. Reference to Fig.8 will show that the heat-treated broken section will really comprise the full wedge shape, as illustrated, thus giving a mechanical joint of some strength.

Bronze castings

Aluminium bronzes occupy a pre-eminent position in many industries because of their mechanical strength and resistance to attack to a wide variety of corrosive media. The duplex aluminium bronzes in the cast form find a wide outlet for pump castings, shafts, valves, and gear wheels, etc.

The welding of aluminium bronze is heavily documented in technical literature. However, there are a few slants on the welding of castings in the repair and maintenance field that are worthy of mention.

The attack caused to component parts working in liquid media can be a complex combination of corrosion, cavitation, pitting, deposit attack and impingement attack, or any single one of these. Figure 10a shows damage caused mainly by cavitation to an aluminium bronze impeller in the paper industry. Figure 10b shows the completed repair. This is interesting from the point of view that Superston-40 weld metal was used to repair this duplex bronze, using the TIG process. An increase in the working life was obtained with an increase in efficiency. No postweld heat treatment was employed.

Stainless steel rings have also been used in this particular area on pump impellers in the mines, but loosening of the rings caused problems. Experimental work was carried out and a successful result obtained by depositing an 18/8/3 Cr-Ni-Mo weld metal to the face. To achieve a weld deposit which would exhibit adequate ductility a thin layer of copper was deposited on the aluminium bronze, using TIG AC, and then a layer of nickel, TIG DC-. The final two layers of austenitic stainless steel were then deposited. When these particular impellers were due for repair, only a deposit of austenitic stainless steel was required. To ensure this the area was initially cut back to allow an adequate deposit of the final layer.

Aluminium castings

Like aluminium bronze, aluminium castings are also widely used and well documented with regard to welding. However, the gear pump casting, Fig.11, was repaired in an interesting way. The worn internal bores were degreased, rotary wire brushed, and the casting preheated to 200°C. Using an oxyacetylene torch and a special flux a tin-based alloy was deposited without fusion. Metallographic examinations have shown that a bond is achieved by the weld metal dissolving the surface of the aluminium, or interalloying. The weld metal used is a derivative of a bearing metal, i.e. tin-based and containing antimony to form SbSn cubes, and copper to form Cu_6 Sn_5 needles. A super wear surface is obtained after boring, and field tests have proven this method.

CONCLUSIONS

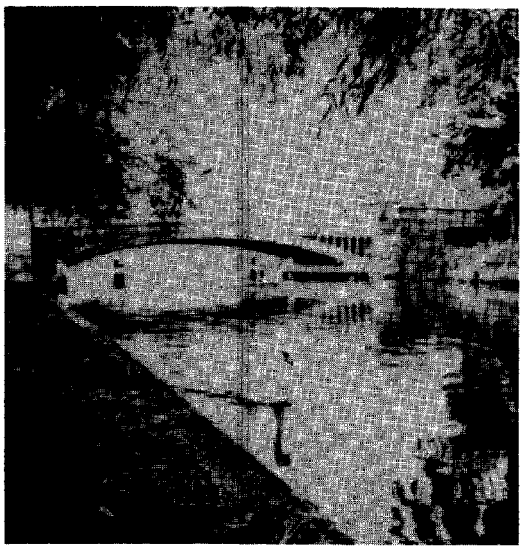
Various selected repairs of castings have been discussed in some detail, even though these repairs do not portray in any depth the full range of problems met in service R and M welding. It is hoped that a general idea has been portrayed to those unfamiliar with this line of repair.

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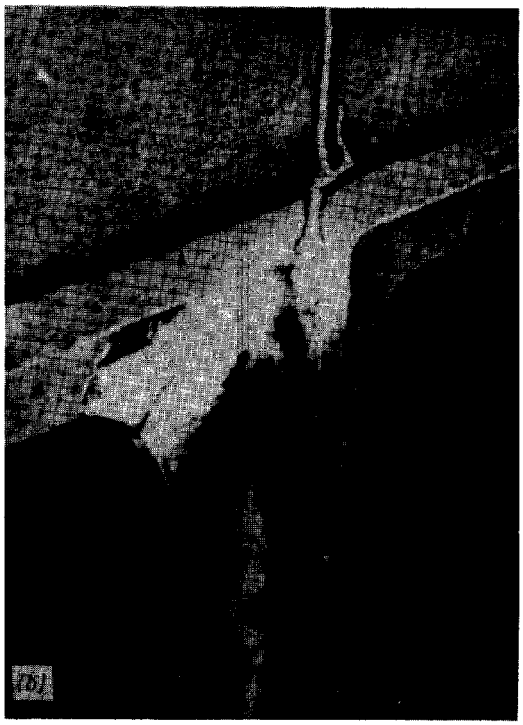
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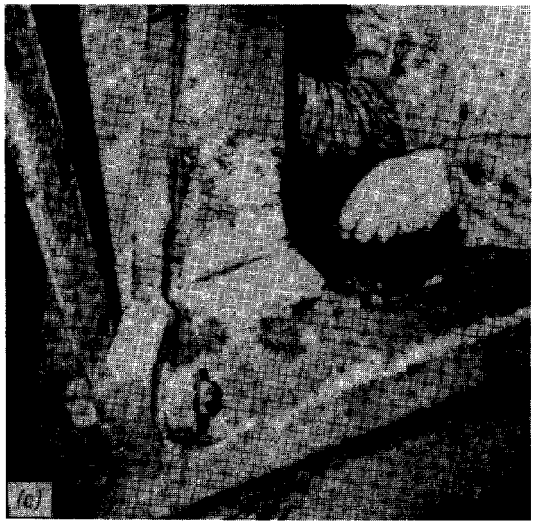
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1 Osney Bridge

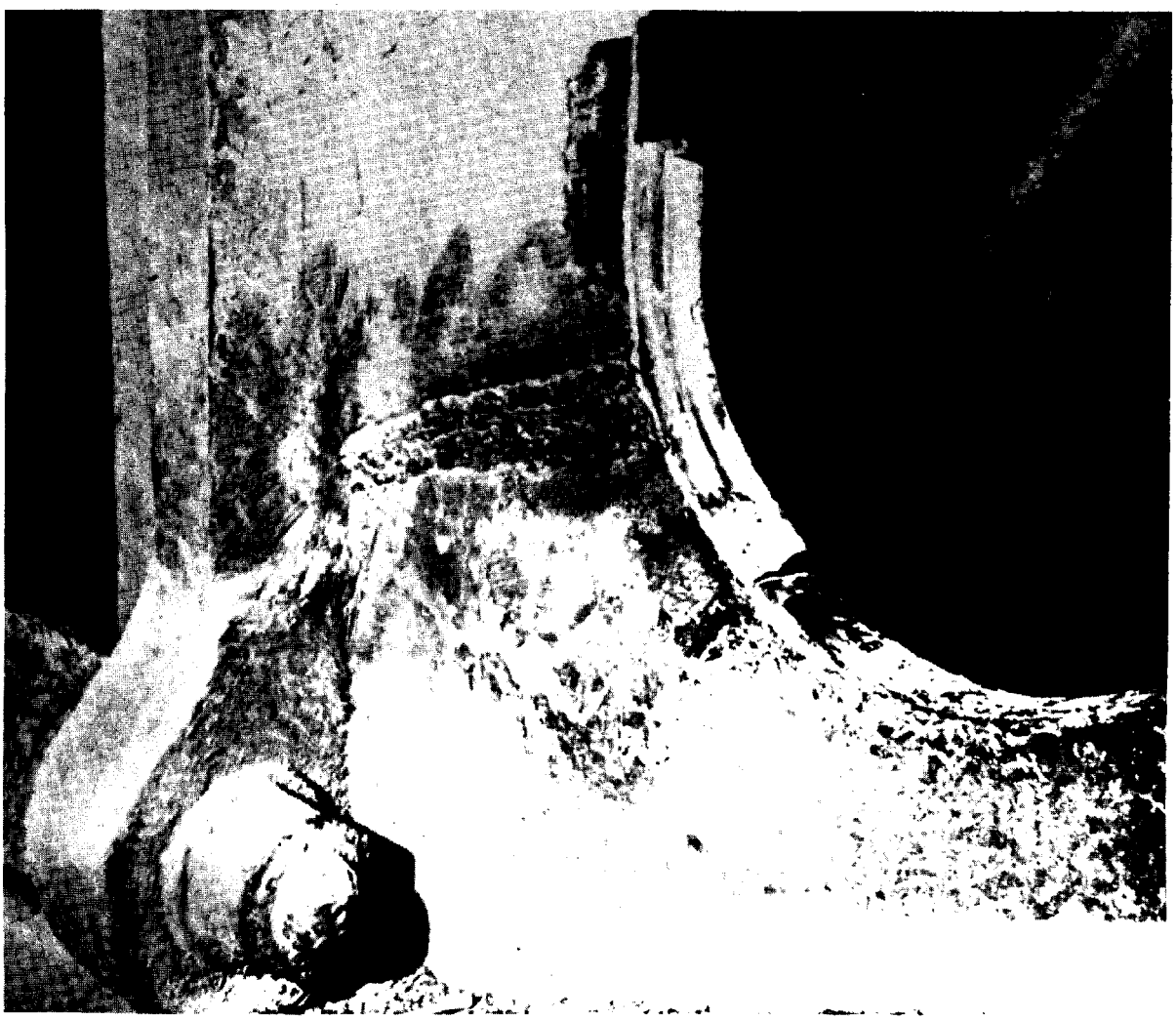




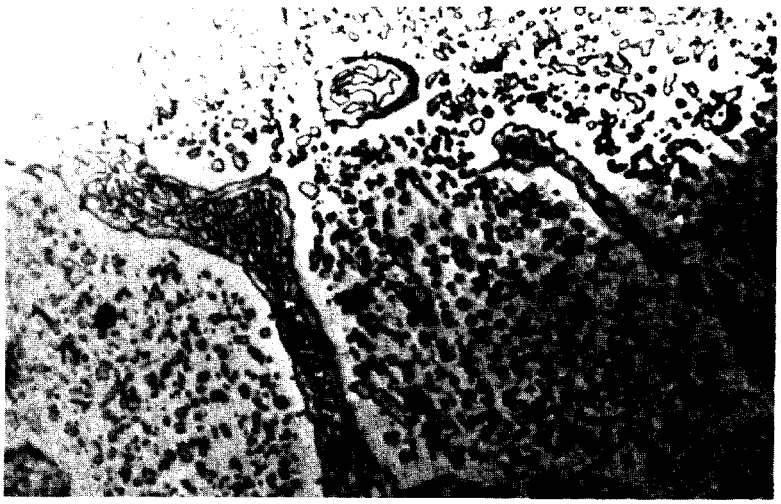


2 Osney Bridge showing: (a) cracks, dye penetrant tested, (b) cutting with hacksaw, and (c) nickel strip in place

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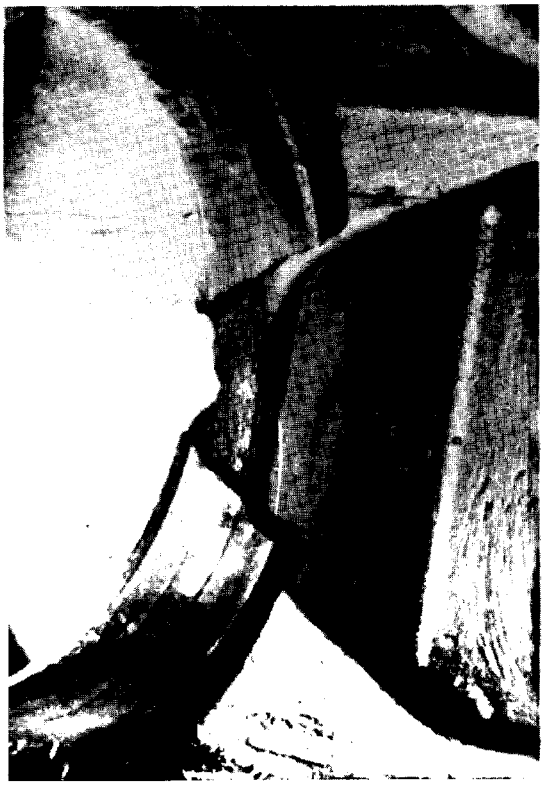


3 Repaired section showing bar in position

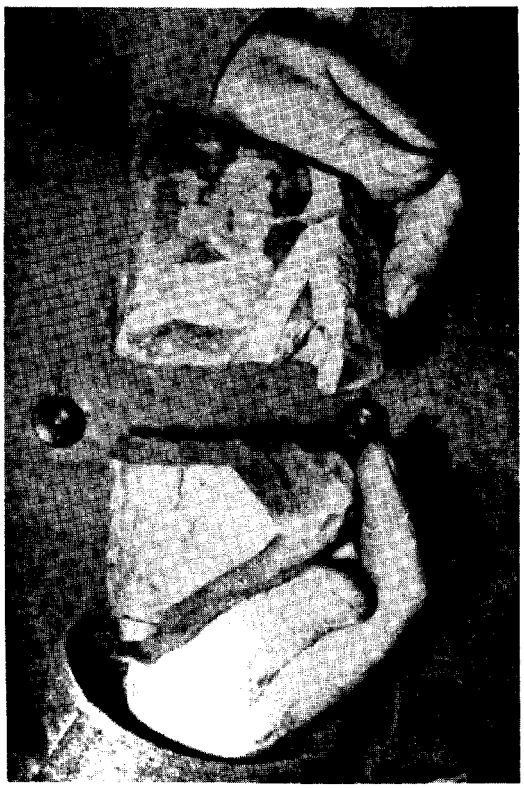


4 Structure of furnace roll x 500

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6 Area of wet liner to be welded



7 Damage to broken-out section



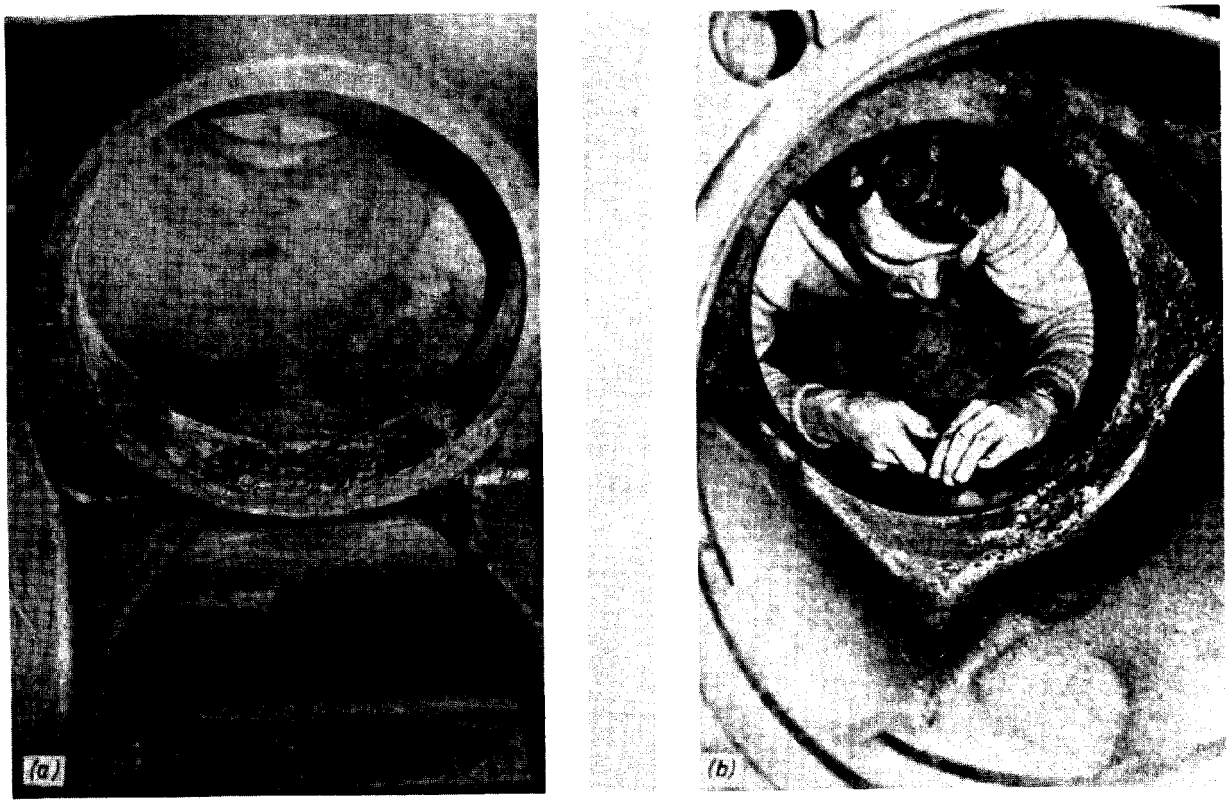


5 Furnace roll showing: (a) clad faces and bars in position, (b) completed roll

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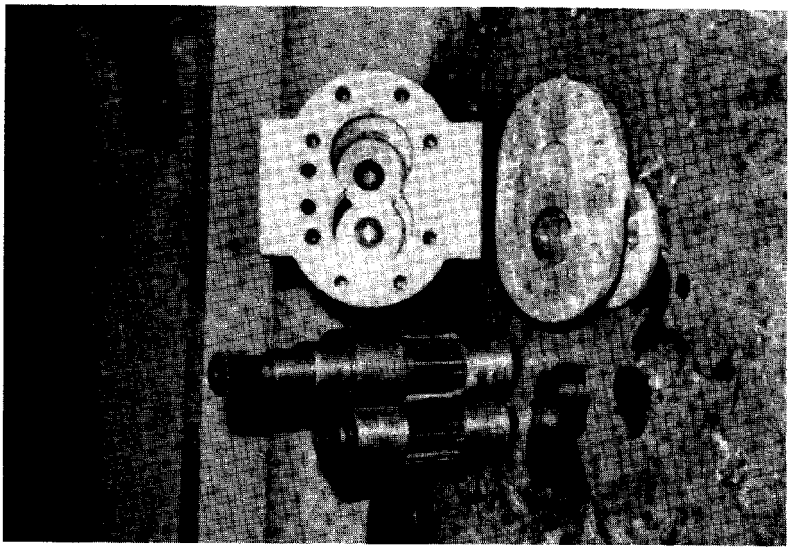


8 Drawing of technique, ...HAZ

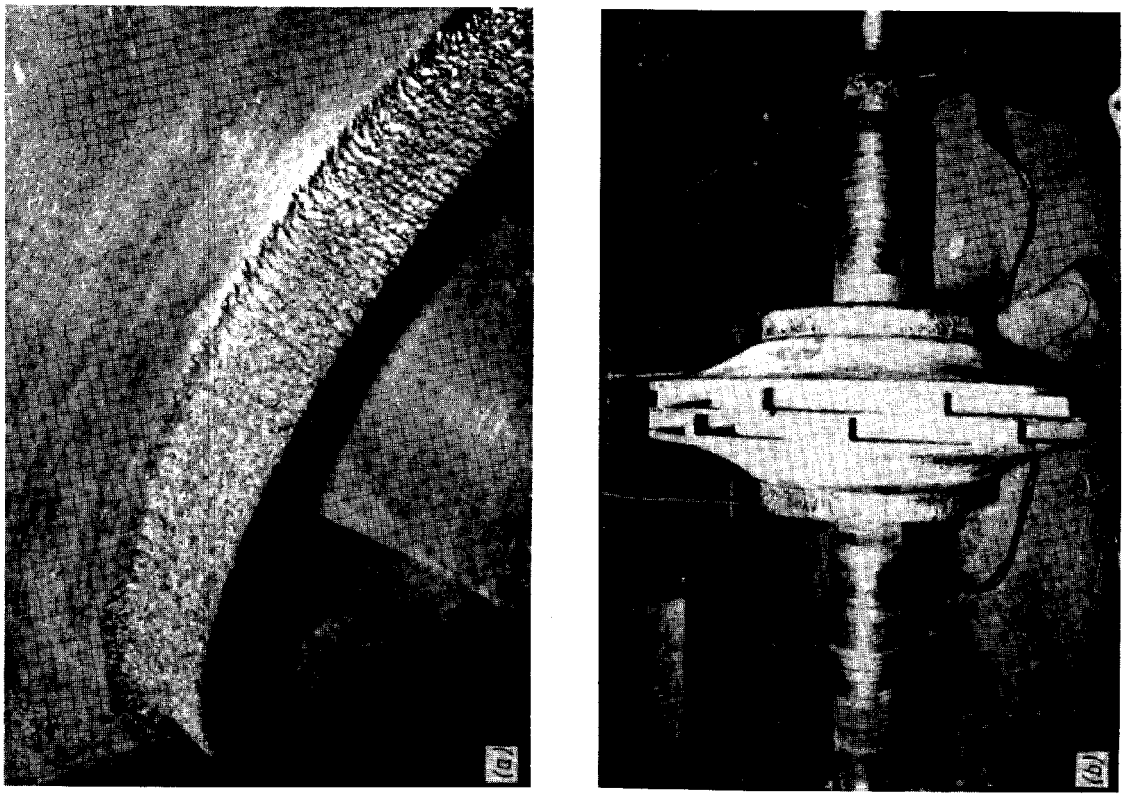


9 (a) completed repair, (b) scraping in operation

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11 Aluminium gear pump casting



10 Aluminium bronze impeller showing: (a) close up of cavitation damage, (b) repaired impeller

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Heat control: key to good cast iron arc welds

by

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THE MODERN, nickel base arc welding electrodes provide a practical method for arc welding many of the cast irons. Today's problems are not in the weld metal, they are in the thin envelope which unavoidably surrounds every weld and which has been heated to a temperature above that at which the carbon (graphitic or combined) goes into solution. It is this carbon solution at temperatures above the so-called critical temperature that makes heat treatment, including hardening and annealing of iron base alloys, practical. It is a fact that while a welder does not intentionally heat treat the material he is welding, he can not avoid doing so, at least in the boundary zone. This means that the key to successful cast iron arc welding lies in control of the cooling rate in the zone wherein the metal was unavoidably heated to above the solution temperature of the carbon. Whenever an iron base alloy is so heated, then the rate of cooling from that temperature controls the hardness of that portion of the metal which was so heated. It is true that the envelope surrounding a weld is very thin, in cast iron, it is rarely over fifteen thousands of an inch thick, but this envelope is critical. It is through this zone that the weld bead fractures and curls up leaving a typical cast iron fractured surface below it. This is because all metals are good conductors of heat and in such cases the heat was very rapidly removed by conduction back into the mass, so much so as to have a quenching effect. This both embrittles and cracks the metal in this thin envelope. The prevention of this undesired cracking lies in control of the cooling rate after the weld is laid down. This involves many factors with mass, ambient temperature, heat input and metal quality, among the more important, as will be seen from a brief look at them.

Mass is important because the greater the amount of metal below the weld the more rapidly is the heat conducted away. Hess and his associates in a classical article published over twenty years ago showed that with all other factors held constant including heat input the rate of cooling in the boundary zone of a weld on one inch thick plate was ten times as fast as for the same weld on quarter inch plate. Thus mass is an important factor in cast iron welding and a technique that would be completely satisfactory on quarter inch-thick material would not be useful on inch-thick stock.

Ambient temperature is important, it is clear that the less the differential the slower is the heat transferred. This too was illustrated by Hess and his data shows that a four hundred degree preheat results in approximately half the rate of cooling in the boundary zone when compared with exactly the weld made with a 72 degree Fahr. ambient temperature. This is an illustrative figure with no magic in four hundred degrees since the less the differential the slower is the heat transfer.

Heat input is also an important factor; in general, the less the heat input the narrower, the more severe and dangerous is the boundary zone with a greater probability for severe cracking. This is due to the fact that the less heat there is to be dissipated the faster does it occur.

PHOTOMICROGRAPH etched to show graphite flakes. Each black wiggle is a flake of graphite (100 diameters).

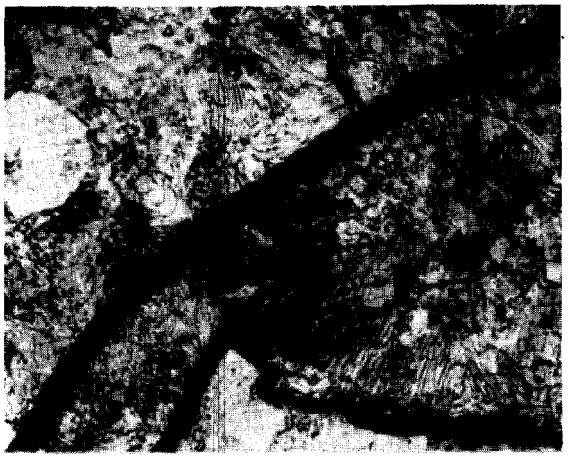


RELATIVELY LOW GRADE iron, etched in 5% Nital. Grey areas are areas containing combined carbon as carbides. (100 diameters).



Source: Canadian Welder and Fabricator, November 1969

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PHOTOMICROGRAPH at 1,000 diameters. Thumb-print-like lines are iron-carbide laminae; iron-carbide is hard and contributes little to material strength.



PHOTOMICROGRAPH of relatively low grade annealed cast iron. (1,000 diameters).

Metal quality is yet another important factor. Every welder is aware that he must use vastly different techniques on high carbon steels than on low carbon steels. A similar situation exists with cast iron. In general the higher the grade of cast iron and the higher its strength, the higher is the proportion of the carbon that is present as carbides. Carbides are much easier to place in solution on heating than is graphitic carbon though the latter does go into solution also, thus the higher the grade (strength) of the iron the easier it is to harden and the more care is required in hardening. Most cast iron welding is a repair operation with little or not data available as to the quality of the iron, in such cases there is only one thing to do, assume that it is high grade iron and act accordingly. If you have low grade iron, no harm has been done but if you assume that your casting is a low strength metal and it should be of the high strength type then you are likely to get an unsuccessful weld so the best plan is to assume high grade metal whenever there is any doubt about it.

In arc welding cast iron it is usual to hold the heat input down in order to avoid excessive contractional stresses. This is generally good practice but we must keep in mind that the cooling rate must be controlled to prevent the excessive hardness that results from rapid cooling. Thus if we are to use low heat inputs and gain the minimum contractional stresses then we must compensate for it by reducing the temperature differential. This means higher preheat. In development work done by the writer he found that on half inch thick, high grade cast iron (45,000 psi iron) with low heat inputs a preheat of 600 degrees Fahr. was essential to prevent any trace of microcracks in the heat affected zone while if it is desired to have a truly machinable boundary zone then the preheat must be raised to 800 degrees Fahr.

Mass is beyond the control of the welder but he must understand it and work with it. He must keep in mind that as metal thickness increases, so does the rate of heat extraction by conduction back into the mass. The curve of heat conduction rates flatten out with increasing thickness and become pretty much of a constant between one and one and a half inch thick material. From there on increasing thickness merely increases the rigidity of the components and this too must be considered when welding heavy sections.

Thus the key to successful arc welding of cast iron lies in the metal immediately adjacent to the weld rather than in the weld itself. The truly critical area is an envelope of less than 1/32 of an inch thick that surrounds the weld and which has been heated above the temperature at which the carbon is dissolved making it available for hardening as the heat is transmitted away from the weld area by conduction. This in turn means that boundary zone cooling rates are a true measure of successful cast iron welding if we may assume good electrodes and techniques.

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IMPROVING THE PROCESS OF GAS WELDING CAST IRON

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SYNOPSIS

It is shown that the properties of cast iron filler wires can be improved by holding in a crucible of a melting furnace at 1400-1580°C. Self-fluxing filler wires alloyed with copper, nickel and tin, and with additions of boron (to 0.03%) and rare earth metals (to 0.12%), have been developed for the welding of cast iron.

On the basis of the type of thermal cycle, the gas welding of cast iron can be divided (provisionally) into three groups (according to the data supplied by VNIIAVTOGENMASh and the Stankolit plant). These are as follows:

1. Welding with the general preheating of components to 600-700°C in heating devices of various types, followed by cooling after welding (under asbestos sheet) from the liquid pool temperature to 400-450°C at a rate of ~0.7°C/sec.
2. Welding with local heating of a defective area to 600-700°C (for instance, with additional portable torches, or directly with the welding torch) and with cooling in air after welding (under asbestos sheet) at a rate of ~3°C/sec.
3. Welding after preheating a defective area to red heat, followed by cleaning the surface by melting and fluxing. After welding, the defective area is covered with asbestos. The rate of cooling of deposited metal in this welding technique is approximately 7°C/sec.

The task of obtaining the required quality of deposited metal at a given thermal cycle is reduced to the manufacture of cast iron filler rods, which make it possible to produce a deposited metal with a structure corresponding to the parent metal and containing no gas pores.

It is recommended that rods with a grey cast iron structure should be used. The formation of cementite of the surface of these rods is unacceptable (1-3).

Investigations show that the nature of droplet formation and of the interaction of the droplet with parent metal is determined by the structure of the cast iron rod (4). On using rods with a ferritic structure and coarse graphite the droplet separates very slowly from the rod (8-10sec) as compared with rods with a chilled cast iron structure or a refined structure (2-3sec). In the first case, a thicker oxide film forms on the droplet, and the spreading of the droplet over the surface of parent metal is poor (despite the fact that the droplet had separated at a higher temperature). Consequently, in the case of using rods with a coarse cast iron structure (cast iron with coarse graphite inclusions), the temperature in the zone of contact between the deposited and the parent metal is considerably higher and the duration of the surfacing process is 1.5-2 times longer than it is with chilled cast iron rods.

During surfacing, the burn-out of carbon and silicon from the rods made of a cast iron with a coarse structure amounts to 22.3 and 20.4% respectively (1.3 and 10.1% for chilled cast iron or for refined cast iron rods). These results can be explained as follows. In the period of droplet formation, an important role is played by the viscosity of liquid cast iron, which, other conditions being equal, is determined by its initial structure. The viscosity of white cast iron at 1300°C is approximately half the viscosity of grey cast iron of the same composition (5,6). This is explained by the presence of graphite blocks undissolved in the iron melt (with a grey initial structure). The process of the dissolving of graphite starts prior to the melting of iron, and it depends on the rate of diffusion of carbon.

Thus, temperature delays occur prior to the separation of the droplet. For cast irons with a coarse structure such a delay takes place at 1450-1500°C, which corresponds to the temperature of dissolving of graphite blocks in the melt (6,8). In this case the duration of the temperature delay prior to the separation of a droplet in these cast irons is approximately three times longer than with chilled cast

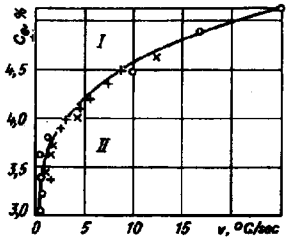


Fig.1. The structural diagram of cast iron: I – grey cast iron (ferrite, pearlite, graphite); II – white and mottled cast irons (pearlite, cementite, graphite).

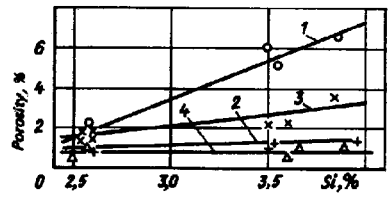


Fig.2. The dependence of the porosity of cast iron deposits on the silicon content of rods and the melting conditions of rods: 1 – overheating 1450°C, 5min; 2 – the same, 1450°C, 20min; 3 – the same 1600°C, 5min; 4 – the same, 1600°C, 20min.

Source: Svarochnoe Proizvodstvo (Welding Production), 23(11), 1976 (Translation published by the British Library)

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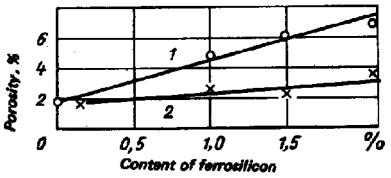


Fig.3. The dependance of the porosity of cast iron deposits on the content of ferrosilicon FS75 in the charge in the production of rods: 1 – overheating at 1450°C, 5min); 2 – the same, at 1600°C, 5min.

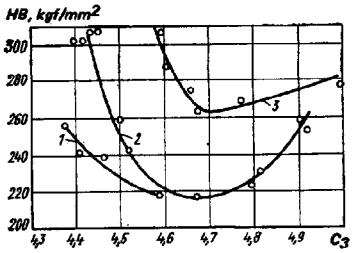


Fig.4. The dependence of the hardness of deposited metal on the carbon equivalent of rods with various C and Si contents: 1, 2, 3 – 4.0-3.6C, 2.6-2.8Si; 2, 3 – 4.0-3.6C, 3.0-3.5Si; 3 – 3;2-3.4C, 3.8-4.6Si.

irons.

Therefore the ledeburitic structure of cast iron results in higher technological properties of the rods, and the deposited cast iron consequently has a more favourable structure and hardness.

Defects in cast iron castings (2,3,7,8-11) are repaired mainly with cast iron filler rods whose composition corresponds to the eutectic condition (C_e = 4.3%). Figure 1 shows the structural diagram of a cast iron constructed in the co-ordinates 'carbon equivalent - the cooling rate v of cast iron'. Analysis of the relationship ($C_e = 3.47\ v^{0.12}$) confirms that at cooling rates typical of gas welding (v = 7°C/sec) the carbon equivalent of the cast iron of the filler rods must not be lower than 4.3-4.4%.

The recommended compositions of cast iron filler rods (GOST 2671-70) are characterised by very wide ranges of the content of elements (for instance, the range of the carbon and silicon contents is 3-4.5 and 2.5-4.5% respectively). But an increase in the silicon content to the upper limit increases the hardness of the metal of the welded joint, as a result of the formation of silico-carbides and alloyed ferrite, and also because of the unavoidable reduction in the contents of total and free (in the form of graphite) carbon. In addition, the susceptibility of deposited cast iron to the formation of pores increases. To ensure that repaired defective areas have the grey cast iron structure and that their properties satisfy the standards for castings of metal-cutting machines, the carbon and silicon contents of the cast iron rods should be 3.3-3.6 and 3.4-3.9% respectively.

In this case it must be taken into account that, during welding, a large amount of silicon is burnt out (up to 20%) at a relatively low burn-out of carbon (5-6%).

The properties of deposited metal are affected also by the conditions of manufacture of rods. A long-term holding (to 15-20min) of cast iron in the crucible of a melting furnace at the melting temperature (1400-1580°C) reduces the viscosity of liquid iron and improves wetting.

Examination of the microstructure of deposited cast iron shows that an increase in melting temperature and in holding time above the optimum values is associated with the de-activation of nuclei of graphite formation, with an excessive refining of graphite, and with the possible formation of supercooled structures in the deposit.

Experience obtained with the application of rods of the cast iron of the proposed composition indicates that efforts to produce a grey cast iron structure in deposited metal are often associated with difficulties caused by the formation of gas pores in the deposit. The principal reason for the formation of pores in deposited cast iron is evidently the presence of hydrogen transferred from the flame of the gas acetylene torch. Under welding conditions, the gas phase contains up to 40% hydrogen, of which 30% can be dissociated into atoms.

The gas bubbles nucleate (at the usual degrees of supersaturation of the metal with the gas being released) on non-wetted inclusion (silicates and aluminates) - liquid metal interfaces.

TABLE 1.

Melt No.	Content of elements*, %										Comment
	C	Si	Mn	S	P	Cr	Ni	Ti	Sn	Cu	
1	3.2-3.6	3.4-4.0	0.4-0.7	0.04	0.15	0.10	0.1-0.2	0.1-0.2	0.3-0.5	-	For repair of defects revealed during machining
2	3.0-3.4	3.0-3.6	0.4-0.7	0.04	0.15	0.10	0.4-0.6	0.1-0.2	-	2.0-3.5	The same, after surface hardening
3	3.0-3.4	3.0-3.6	0.4-0.7	0.04	0.15	0.10	0.2-0.5	0.1-0.2	0.1-0.3	1.0-2.0	The same, on scraped surfaces

* In all cases the contents of hydrogen, oxygen, sulphur, phosphorus 0.0065, 0.04, 0.15 and 0.10% respectively, and the contents of manganese and titanium should be within the range 0.4-0.7 and 0.1-0.2% respectively.

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TABLE 2.

Filler material (content of elements, %)	Flux	Characteristics of deposited cast iron		
		Machinability	Porosity, %	Hardness HB, kgf/mm^2
Cast iron (3.42C, 3.5Si, 0.63Mn, 0.2P, 0.05S, 0.45Ni, 1.52Cu)	Without flux	Good	0.231	224
	Borax	Good	0.060	220
Cast iron (the same + 0.03B)	Without flux	Good	0.537	236
	Sodium carbonate	Good	0.150	230
Cast iron (the same + 0.2B)	Without flux	Satisfactory	0.032	264
	Sodium carbonate	Satisfactory	0.017	289
Cast iron (the same + 0.05Ce and 0.03Y)	Without flux	Good	1.812	236
	Boric acid	Good	0.120	229
Cast iron (the same + 0.07Ce and 0.05Y)	Without flux	Satisfactory	4.9	235
	Boric acid	Satisfactory	2.3	248
Cast iron (3.44C, 3.62Si, 0.45Mn, 0.13P, 0.06S, 0.48Ni, 2.26Cu, 0.02B, 0.08Ce, 0.04Y)	Without flux	Good	None	202

The risk or pore formation can be reduced by lowering the content of various non-metallic inclusions in the cast iron, by selecting a suitable chemical composition of the cast iron for rods, and by adhering to the optimum casting conditions.

The investigations conducted at the Stankolit plant, at the Moscow Institute of Steel and Alloys, and at VNIIAVTOGENMASh (12), show that an increase in the silicon content of rods increases the susceptibility of deposited cast iron to the formation of pores (see Fig. 2), especially with an increasing fraction of ferrosilicon in the hot metal charge in the melting of iron for rods (see Fig. 3). A relationship has been determined between the formation of pores and the number and nature of non-metallic inclusions in the cast iron of the rods. The composition of inclusions, and their nature and dimensions, determine the overall adsorptive effect. Non-wettable silicates are the most 'dangerous', because they increase the interfacial area, and consequently reduce the work of formation of a gap pore. An increase in melting temperature, and particularly in holding time in the furnace, in the melting of cast iron for rods facilitates a more complete removal of gases and non-metallic inclusions from the cast iron, and the porosity of cast iron in repaired areas in castings consequently decreases. In the melting of cast iron in induction furnaces, the optimum melting temperature (1400-1580°C) depends on specific conditions. The optimum time of holding the cast iron in the furnace prior to pouring into moulds is 15-20min.

The predetermined ratio of carbon to silicon in cast iron, and the conditions of melting in the manufacture of rods, made it possible to produce poreless deposits. But the strength and wear resistance of the welded joint decreased (due to the presence of up to 70% of ferrite in the joint).

The results of investigations into the effect of various alloying elements, added to the rods, on the properties of deposited cast iron show that completely poreless deposits, with the structure and hardness corresponding to that of grey cast iron, can be produced as a result of alloying with nickel, tin and titanium, with nickel, tin and copper and with nickel, titanium, tin and copper* (see Table 1).

In the welding of cast iron, the flux is usually represented by borax, which decomposes at high temperatures to boron anhydride and sodium oxide. The sodium oxide facilitates binding of the oxides (mainly FeO and SiO_2) present in the cast iron, and also those formed during welding, in complex compounds. The presence of boron leads to the formation of low-melting compounds in the form of a dense film on the surface of the weld pool. But a uniform supply of flux is required to produce a continuous slag cover, and this cannot be achieved even by using various devices. Consequently, the deposited metal often contains a large number of pores.

A continuous slag cover can be formed with self-fluxing cast iron rods containing boron and elements that form basic type oxides with a low dissociation tension (e.g. cerium and yttrium).

The rods with up to 0.2%B and up to 0.7% of rare-earth metals (REMs) can be used for the welding

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of cast iron without the need to add any additional quantities of flux to the weld pool.

To clarify the effect of alloying elements on the technological properties of the welding process, experimental deposits were made with and without flux. In deposition with cast iron rods containing boron the flux was represented by sodium carbonate (Na_2CO_3), and borid acid was used with rods of a cast iron alloyed with rare-earth metals. Both boron and the REMs, added to the cast iron of the rods, exert the fluxing effect during welding (see Table 2). This reduces the porosity of cast iron deposits. The burn-out of boron in welding varied from 30 to 65%.

Increasing the REM content of the rods increased the porosity of cast iron deposits (this increase was especially pronounced at REM contents of more than 0.3%). The machinability of deposited metal was impaired also. The higher hardness of deposited cast iron, as compared with the parent metal, is obviously due to the greater burn-out (to 30%) of silicon during welding.

Experiments in which boron, cerium and yttrium were added separately to the cast iron of the rod (see Table 2) gave the following results: the addition of boron facilities the formation of a dense film of slags on the pool surface, but the characteristic feature of the process is a high viscosity of the metal of the pool. This extends the duration of deposition, and it increases the burn-out of silicon; the hardness of cast iron deposits consequently increases until chilled cast iron structures are formed; the addition of cerium and yttrium satisfactorily deoxidises the metal being deposited, but it increases the porosity of the deposit.

The combined addition of boron, cerium and yttrium (at a specific ratio) increases the quality of welded joints, and it creates favourable conditions for the self-fluxing of filler material in the course of welding.

The highest properties of cast iron deposits were obtained with rods with carbon equivalents of C_e = 4.6-4.8% and with 2.6-3.5%Si (see Fig. 4).

At silicon contents in excess of the above limit, the hardness of the cast iron deposits increased to a considerable extent as a result of the formation of silicocarbides. Annealing at a temperature of 500-600°C, corresponding to the decomposition temperature of the silicocarbides, reduced the hardness of the deposits.

CONCLUSIONS

1. It has been established that holding in the crucible of a melting furnace at 1400-1580°C has a beneficial effect on the properties of cast iron filler rods.
2. Self-fluxing filler rods, alloyed with copper, nickel and tin, and with additions of boron (to 0.03%) and REMs (to 0.12%), have been developed for the welding of cast iron.

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QUENCH WELDING PROCESS FOR JOINING CAST IRON

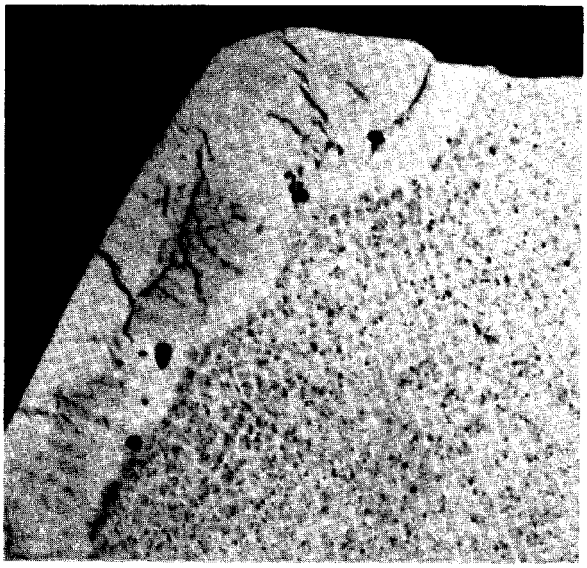
CLIVE COOKSON*, M.Weld.I.

THE metal-arc welding of cold cast iron has for many years been a controversial subject, mainly because of the high failure rate both during and after welding. The development of specialized electrodes and welding techniques has more recently established this process for the repair, reclamation and production welding of cast iron.

The Quench Welding procedure introduced in March 1960, after three years of field testing, is an even more radical departure from the full fusion oxy-acetylene process because of the integral quench known as Step C in the joining operation. Failures during and after welding can be caused by:

1. The heat-affected zone beneath the weldment, which consists primarily of eutectiferous carbides, martensite and bainite. This forms a metallurgical notch, and allows the development of stress in a straight line.
2. The low ductility of the flake type iron which comprises the larger proportion of castings, and the inability of this iron to absorb stresses created by base metal expansion and weld metal contraction.
3. Incorrect weld metal chemistry of the ferritic type on large fills or in conditions involving restraint. Ferritic deposits on cast iron virtually become hardened tool steel due to carbon pick-up and the quench rate owing to the cold base metal mass. Such deposits are susceptible to quench cracking and hydrogen induced cracking in the heat-affected zone under the usual welding conditions. Fig. 1 shows a ferritic deposit made as a buttering pass to achieve the optimum conditions by eliminating a reaction stress. The visible fine cracks would probably go unobserved, and failure could well occur during service.
4. Limited knowledge of preparation, pre-weld testing, welding and peening techniques, and the ability to recognize the arcing characteristics of the nickel type deposit.
5. High heat input causing deleterious expansion and base/weld metal dilution.

Fig. 1. Ferritic type deposit on grey cast iron. The surface has only been buttered but extensive cracking is visible.



The formation of a "white cast iron" heat-affected zone cannot be entirely avoided during the metal-arc welding of cold cast iron, but it can be reduced by using low heat input electrodes containing caesium, strontium, barium, etc., to impart high thermionic emissivity at arcing temperature levels. The increased ionization across the arc gap is instrumental in reducing the operating amperage and the total heat input. The metallurgical notch, and stress in a straight line, can be reduced by making this area discontinuous.

The non-ductility of grey cast iron can be offset by using weld deposits which have a low hot yield strength so that stresses can be absorbed by the weld metal without cracking. Peening also stretches the weld bead and changes the tensile stress to a more desirable compressive stress.

Pre-weld testing is as important as the preparation, and also the ability to use a nickel-type electrode at the correct amperage. Suitable deposition techniques are also essential in achieving a successful repair.

The cheap ferritic deposit can be employed where a non-machineable weld is permissible. The deposit has the advantage of rusting and colour matching the cast iron. However, the high yield point developed by carbon pick-up from the casting makes this deposit type totally unsuitable where stresses must be absorbed during welding.

Throughout the welding operation it is essential that the base metal temperature should not exceed 80°C. in the area of welding. To ensure this, sufficient interpass pauses are required. Cast iron has a higher thermal conductivity than mild steel because of the graphite content. Its expansion is approximately 0·000012 in./°C. A surface temperature of 1,000°C. will cause an expansion and contraction of 0·012 in. This results in a stress concentration which can lead to cracking. Many castings are unweldable. The Quench Welding process overcomes these two problems, and considerably reduces welding time.

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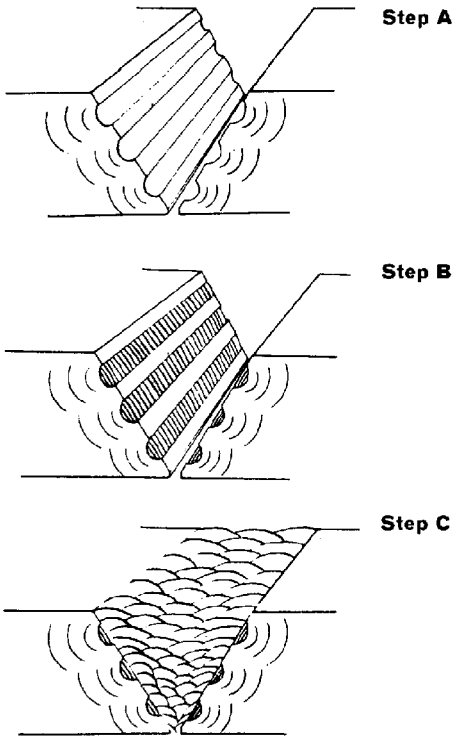


Fig. 2. The three main stages in the Quench Weld method of repairing cast iron. Step A: grooving operation. Step B: filling of grooves. Step C: joining operation.

The Tab Test

Before an attempt to repair is made, the casting should be tested for weldability. The tab test is a method used to ascertain this, and not to determine the type of cast iron. A hot-rolled piece of mild steel, ¼ in. by 1 in. by 1½ in. should be double fillet welded to the casting near the joint area, using electrode QuenchTrode C. When the weld has cooled to room temperature the steel strip should be hammered off. A weld that breaks off flush with the surface denotes impurities in the base metal.

The next step should be spraying across the surface with the iron-base electrode, QuenchTrode B. A second plate should then be double fillet welded into position. If this breaks off flush with the surface then the casting must be considered unweldable. This situation, however, is rare.

Where a casting has broken due to overloading, a weld repair would be subject to failure. Reinforcement of the part with strongbacks or gussets is usually a solution, but the load limits and design factor must be evaluated by a qualified engineer to avoid the development of a dangerous situation.

Grey iron does not exhibit a definite yield like mild steel, though at room temperature it can sustain loads of up to 80% of its tensile strength. However, it is usually designed for service at about 25% of its tensile properties.

Sulphur and phosphorus are elements which can lead to weld failure if they exist in amounts in excess of the safe limit. Sulphur produces excess cross bead cracking and the pick-up of as little as 0·1% by weight in a high nickel weld deposit will cause brittle fracture. The weldability of a casting is affected when the sulphur content exceeds 0·15%.

The maximum phosphorus content in cast iron which allows freedom from cracking is 0·6%. Where this figure

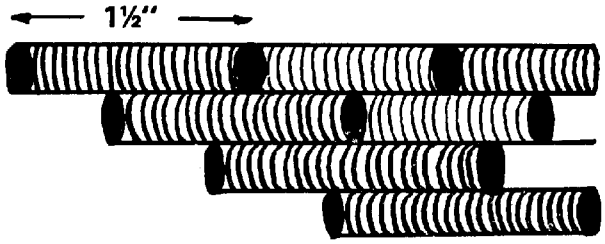


Fig. 3. Alternate deposit sequence used when filling a V joint. The black dots indicate where craters have been filled.

is exceeded, the amount of the iron phosphide eutectic round the grain boundaries is pronounced. This constituent melts at about 960°C, and under the influence of stress, intergranular cracking can occur due to the chain formation of the iron phosphide which is hot short when the solidified weld bead is contracting.

Nickel has a slow solidification rate and this allows time for the formation of gases from sulphur or base metal contamination by oil, paint, or grease, and the deposit is repelled. An iron-base electrode which has a high solidification rate overcomes this problem by sealing in the surface before the final joining operation. The electrode, QuenchTrode B, should not be used for joining.

Preparation

Oil, grease, rust, paint, etc., must be completely removed from the weld area. Ferrous hydroxide, which forms part of the corrosion product, decomposes to steam under the arc which can lead to heat-affected zone cracking.

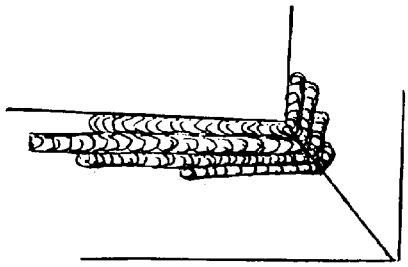
Damage to a casting may take the form of:

(a) An incomplete crack.
(b) A complete crack separating the component.
(c) A situation involving both the above conditions.

In all repairs, the surface must be cleaned and then dye penetrant tested. The cracks should be centre punched along their length to make a blueprint of the damage, and then holes drilled $\frac{1}{16}$ in. in advance of the cracks to allow for sub-surface extension. The size of the drilled hole will be determined by the thickness of the section, though $\frac{1}{8}$ in. usually suffices.

Double V preparation should always be employed when possible for obvious reasons, and where this is not possible a single V preparation, two-thirds thickness of the material will usually suffice because of the high strength of the deposit and the surface area covered in the joint design.

Fig. 4. Corner weld extension used to prevent edge cracking.



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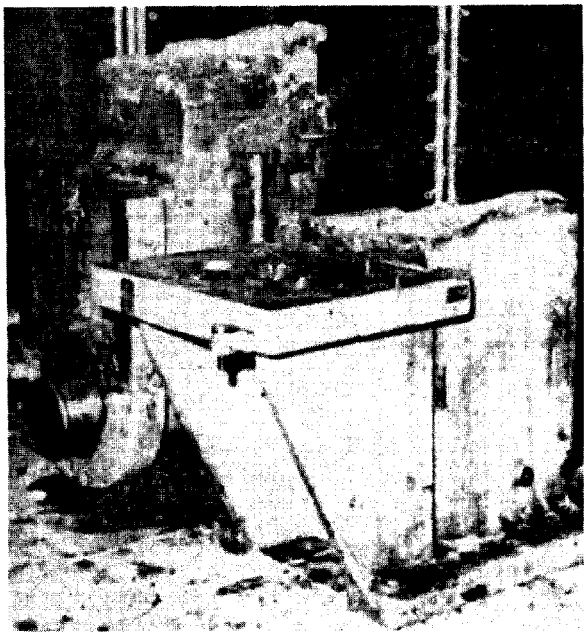


Fig. 5. (Above) Broken parts of a 40 ton press with stress grooves prepared.



Fig. 6. (Right) Welding together the broken press frame.

Sharp corners must be radiused to prevent dilution and the formation of hard spots. A rotary cutter, unlike the grindstone, does not shake the free graphite to the surface, though it is advisable to wire brush the prepared area vigorously before welding.

Quench Weld Step A. Step A (Fig. 2) is the grooving operation, and may be accomplished with ChamferTrode, an oxygen-free metal-removing electrode which has a specially thick coating so as to concentrate the arc to give the maximum blasting effect.

Once the joint has been prepared either by cutters, grinders, ChamferTrode, or a combination of these three, stress-relieving grooves must be cut across the joint faces. The grooves should be three-quarters the diameter of the coated ChamferTrode, and spaced not closer than $\frac{1}{4}$ in. apart. The maximum size ChamferTrode employed for the grooving operation is $\frac{1}{8}$ in. diameter.

Step B. The next operation is the filling of these grooves, using the iron-base electrode which has a high solidification rate and is designed for greater cleansing properties. The back-step technique should be used, and each deposit soundly peened. No quenching should be attempted at this stage, and the deposits must be allowed to cool naturally. The bead should be flat with the surface, and only a single pass made to fill the groove.

Once the grooves have been filled, QuenchTrode B should again be used. Drawing a long arc, the areas between the deposits should be sprayed, again allowing the area to cool between each spray. The long arc reduces the heat input, achieves a 0·020 in. deposit thickness, and effectively seals in base metal contaminants which could be responsible for poor bonding and porosity caused by outgassing.

Step C is the final joining operation in which an air jet or a damp rag is utilized. The rag should be squeezed free of water and pressed on the deposit immediately the arc is broken. When an air line is used, peening and quenching can be done simultaneously, and so an air blast offers advantages over the damp rag quench which must be closely controlled to avoid the presence of water in the joint during welding, in fact this technique should be used only as a last resort.

The deposit length should be a maximum of 4 in. and this should be immediately peened along the bead, from the crater end, with blows sufficient to indent and slightly spread the surface.

An arcing block is a useful aid because this allows the electrode tip to reach its operating temperature before being transferred to the job, and so minimizing the heat input to the base metal.

Craters are dangerous crack starters, and these can be avoided by pausing at the end of the run for 3 s to allow a minute accumulation of weld metal.

When filling the joint, the stringer beads should be alternated (Fig. 3) so that an unsightly accumulation of lumps does not occur at one end of the joint. Stresses, too, are minimized by finishing the subsequent run approximately in the centre of the previous deposit.

Weaving should not be employed during the welding operation because this increases the thermal gradient for the same length increment. EutecTrode C is formulated for wash and fluidity. Wash is the lateral capillary flow of the deposit, and fluidity the forward flow.

The short arc brazing technique is a method where the arc is concentrated on the molten metal which flows forward and so prevents the arc from "digging in". Progression of the deposit should be co-ordinated with the intervalled electrode burn-off rate, a characteristic inherent with the nickel type deposit. A higher amperage alters the characteristic to that of a mild steel deposit, and this explains one reason for the failure of a component where a welder is not familiar with this aspect of the work.

The corner weld extension (Fig. 4) is a vital part of the correct technique. Cracking is avoided by taking the weld

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¼ in. round the edge of the joint, so that it contracts on to the vulnerable section like a "crooked" finger.

When the joint has been completely finished, grooves should be chamfered across the surface 1¼ in. apart. The length of these lacing beads will depend on the width of the weld, but in any event these should extend each side by ¾ in. and 1½ in. The "long" and "short" beads should be alternated around the joint. It is essential not to peen these lacing beads so that contraction of the deposit has a bracing effect. Porosity is caused by:

(a) The presence of water in the joint when quenching with a wet rag, a reason for substituting this by an air line.
(b) Moisture pick-up in the electrode coating which is cured by baking at 150°C for 3 h prior to use.
(c) A rapid rate of travel which disrupts the gaseous shield of the arc.

Cross checking is the cause of a high heat input, (i.e., excessive amperage) or weaving. Some slag types are difficult to refloat when weaving, and this can result in slag inclusions or porosity due to the trapped slag gasifying.

Hot peening will alter a high and narrow convex bead, which is crack sensitive, to a more suitable shape, and cold peening will remove stresses in the weld bead.

The high strength irons (Meehanite, etc.) can be joined with a strontium-aluminium bearing electrode (25 ton/in²), containing degasifiers in the coating to enhance the welding operation, and alloying elements to refine the grain structure and provide for an accelerated cooling rate.

Metallurgical Aspects

The metallurgy of the casting design of the component and the cooling rate during the casting process are important factors, and it will be found that no two castings are identical from a welding point of view.

A lack of balance in the cooling rate and the resistance of the mould to the contraction of the solidified metal during casting can cause residual stresses, some of which have been reported as high as 14 ton/in².

Though the various stress patterns will balance out during cooling, these can be disturbed by metal removal, i.e., machining or preferential heating. The grooving in Step A helps to disperse stresses in and near the joint area. (Figs. 5 and 6).

Manipulative Skill

It is essential that the welder should understand the various techniques of welding cast iron and that he should be fully competent to carry them out. The welder is certainly the most important man in the repair operation because success or failure rests entirely on his skill.

The Quench Weld process should not be used on chilled irons (i.e., White cast iron, Ni-Hard, etc) which are inherently crack sensitive under ordinary conditions. This base metal would crack during the quenching operation.

The Quenching Process

The experts were initially dubious about the idea of quenching a hot weldment and base metal with an air line or damp rag, the latter being significantly prone to hardening and cracking.

However, if one considers an initial 2 in. deposit length of an electrode on a casting of ⅜ in. minimum thickness, upon breaking the arc the temperature will be a maximum of 110°C and be falling rapidly. An air line or damp rag is now used to accelerate the cooling to approximately 20°C. The second and subsequent deposits are the same as the first and so the operation continues identically, i.e., always quenching at 80—110°C.

The Ms—Mf range of cast iron depends on the chemical composition and alloying elements. However, Ms 220°C and Mf 90°C is the general range for grey, spheroidal graphite, and malleable irons when unalloyed.

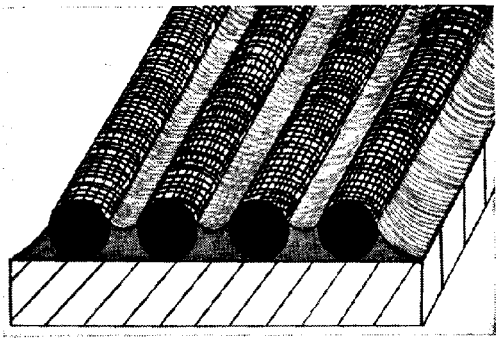


Fig. 7. The CastoGuna process. Profile of welds between reinforcement bars.

The wideness of this quoted range can be affected in unalloyed irons by the silicon content which, when decreased acts similarly with the Ms point.

The austenitizing temperature also influences the Ms point which again is depressed where high temperatures are encountered, and so causing more carbon to dissolve in the austenite.

However, the important fact to note is that the Mf temperature is never below 90°C, no additional hardening can occur from quenching from a maximum of 110°C.

The benefits of using the Quench Weld process with specially formulated alloys are considerable because the main source of cracking is totally removed, i.e., pronounced base metal expansion and contraction.

The requirements of a suitable weld deposit for cast iron are, therefore, numerous, and they include:

(a) Full machinability.
(b) Low yield point.
(c) High ductility which allows substantial peening.
(d) Low heat input.
(e) Arc stability.
(f) Instant arc ignition.
(g) Response to accelerated cooling, etc.
(h) Low hot-short strength.

The CastoGuna Technique

The CastoGuna welding technique was developed by Von Saurma, a German Castolin-Eutectic welding engineer, and depends upon the reinforcement of a weld area with steel rods (Fig. 7).

Its main use is for the repair of heavy parts which might be prone to weld metal contraction stresses where large deposits of weld metal are required. It was first used in England at a large steel works on the cast iron frame of a hot strip rolling mill which was fractured through a shoulder (Fig. 8). The component weighed approximately 50 ton. On some repairs, the CastoGuna process can be used across the fracture without first welding the joint. The reinforcement of the weld area is obtained by welding round bars transverse to the weld or fracture. Important points are:

1. The bar material should be EN8 or similar material quenched but not tempered to obtain a high yield point.
2. The rods, cut to the same length, are placed across the weld or fracture, and they should alternate off centre to obtain an overlap not less than 4 in. from the edge of the weld on the short side and 6 in. on the long side. The bars will, therefore, measure 10 in.

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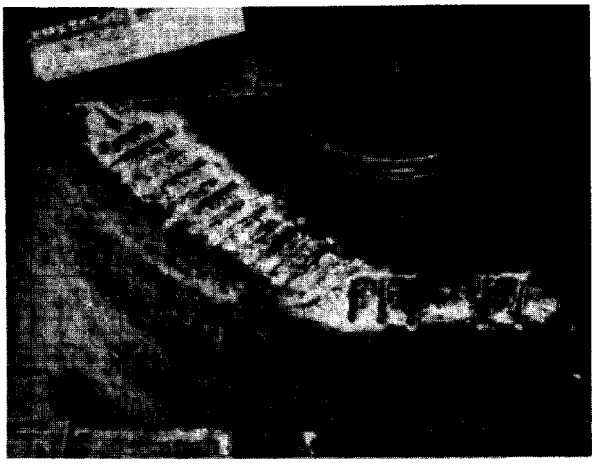


Fig. 8. Fractured steel mill frame showing reinforcement bars welded in position.



Fig. 10. Repair by the CastoGuna method of a cast iron pulley wheel at a brick works. The illustration shows one reinforcement in position.

together with the width of the surface weld or 10 in. when pre-weld jointing is not observed.

3. Round bars are used because good root penetration is obtained leaving a small area of unwelded base metal.
4. The spacing between the rods should be one and a half times the diameter of the rods, and the area between should be filled with weld metal to achieve a complete band round the fracture. The coverage of the surface with weld metal is more important than the thickness of this coverage or reinforcement.

When the bars have been marked in position with chalk lines, stress grooves are cut, ¼ in. apart with ChamferTrode at right-angles to the bars. The number of stress grooves (Fig. 9) should be a minimum of five on the 4 in side of the bar, and seven on the 6 in. side. These grooves are then filled using an iron-base electrode, QuenchTrode B. The grooves should extend ⅜ in. on either side of the bars.

It can be seen that when these grooves are filled, and the bars welded into position with the weld reinforcement between them, the "grip" to the base metal will be extensive and conclusive, acting as a series of dovetail joints.

Welding of the bars into position should be carried out with QuenchTrode C, and the air line quenching method employed. The deposition method is extremely important, i.e., welding should begin at the top of each rod and a 1—1½ in. bead deposited. Three superimposed fillet deposits should be made either side before the next 1—1½ in. is tackled, working down the bar. It is usually advisable to complete the welding on one end of all the rods before moving to the opposite end.

After the rods have been completely welded in this manner the between spaces are then filled, forming a concave shape, each end reaching the centre of the bar. Any additional weld metal will not increase the strength.

The techniques regarding peening and crater filling are the same as those used in the Quench Weld process, and should be observed.

The bars can be heated and bent to any configuration, and then reheated and quenched. The diameter of the bars will depend upon the section thickness of the casting in the area of fracture, but they are seldom less than ¾ in. or more than 1¼ in. Where pre-weld joining is omitted, the strength of the reinforcement becomes singularly important.

Fig. 9. Reinforcement bar welded in position with stress grooves filled below it.

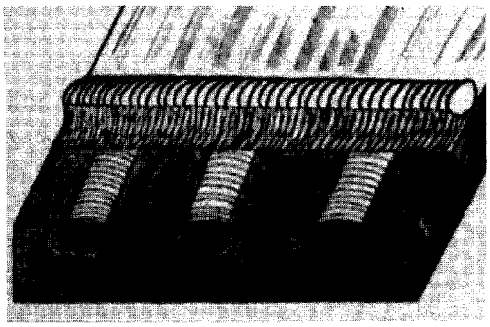
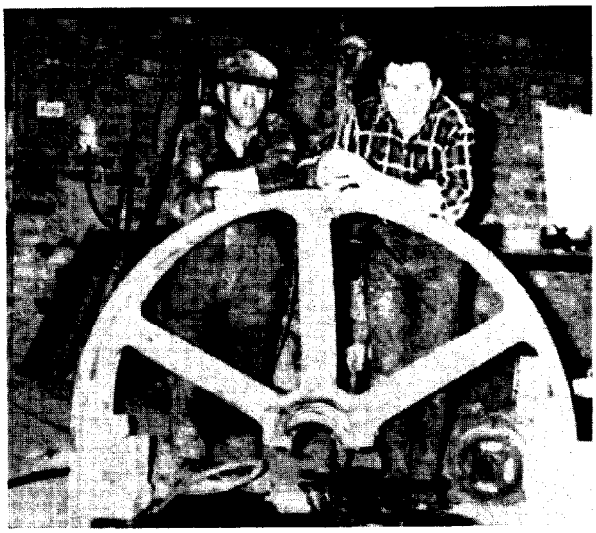


Fig. 11. Completed repair on the pulley wheel.



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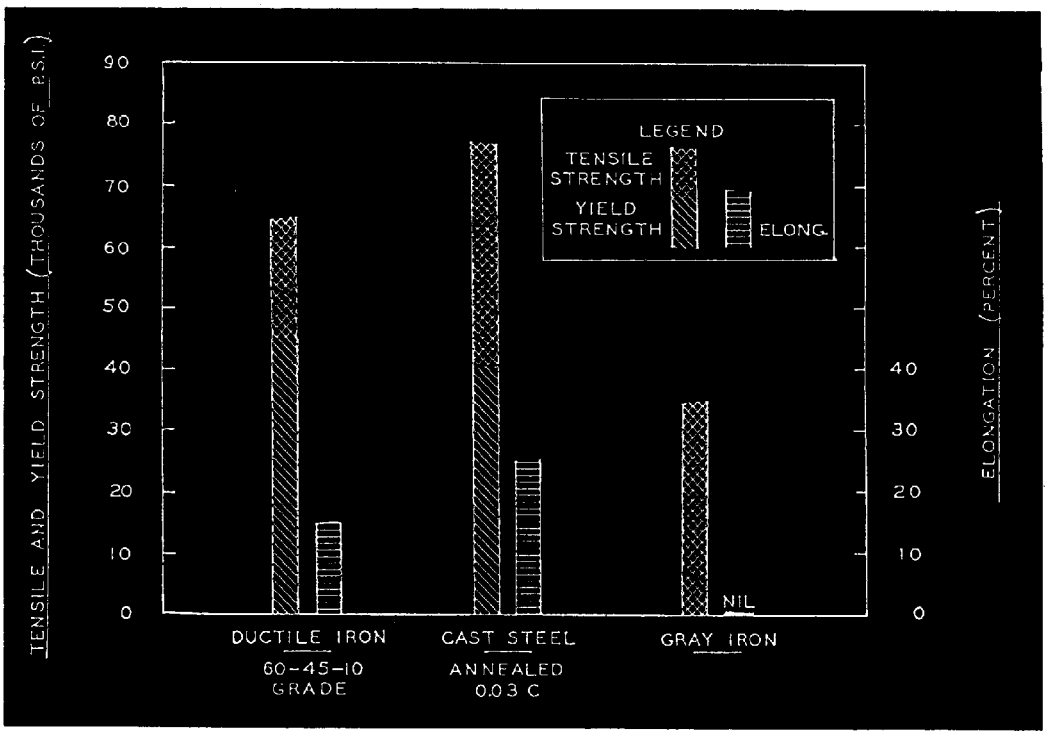


FIG. 1 — Mechanical properties of some cast products (unwelded).

Welding ductile iron— Part One

by J. W. Flannery
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LATEST DEVELOPMENT in ductile iron welding is use of the short circuiting arc process. Because of its lower heat input, this process can be used to improve mechanical properties and machinability of ductile iron joints. Although ductile iron is being welded by this process commercially in Europe, its use on ductile iron in this country is just beginning to be evaluated.

Ductile iron has become a widely accepted engineering material since its introduction 20 years ago. This material is also known as nodular iron or spheroidal graphite iron. In the western hemisphere the name ductile iron is used, while in Europe and most other areas the term spheroidal graphite iron, or S.G. iron, is preferred.

Like gray irons, ductile irons are actually a family of cast alloys. Based on microstructure, there are five main types of ductile iron: pearlitic, ferritic, acicular, martensitic and austenitic. The first four are similar in composition to ordinary gray irons but are more ductile and have tensile strengths two to three times that of gray iron.

With respect to mechanical properties, ductile iron is often likened to cast low carbon steel. Ductile iron has been referred to as "the cast iron that can be bent."

The substantial improvement in properties over those of gray iron is made possible by the introduction of an appropriate quantity of magnesium into a heat of gray iron followed by inoculation with ferrosilicon. The effect of this treatment is to replace strength-impairing flakes of graphite with relatively harmless spheroids.

The various grades of commercial interest have been classified by ASTM according to their mechanical properties (Table I). In most of the specifications, the agreed upon mechanical properties may be obtained by use of suitable combinations of composition and heat treatment, at the option of the supplier. Depending on composition, the as-cast elongation of ductile iron usually falls within the range 2-20%. In most cases, ductility can be improved by the ferritize-annealing heat treatment shown in Table III. By this treatment, both primary carbide and pearlite are transformed into spheroidal graphite and ferrite.

Weldability. Most of the published welding work has been on material which received this "ferritize-anneal" heat treatment. Very little welding has been done on the high-strength irons. Some work has been performed on austenitic compositions.

The conclusions which are derived from an evaluation of weldability are tied not only to the method of evaluation and the test results, but also to the particular criteria of weldability selected. In some situations, an alloy must yield welded joints having properties equivalent to those of the unwelded material. In others, it may be satisfactory to obtain weldments adequate for intended service.

In general, ordinary flake-type gray iron has

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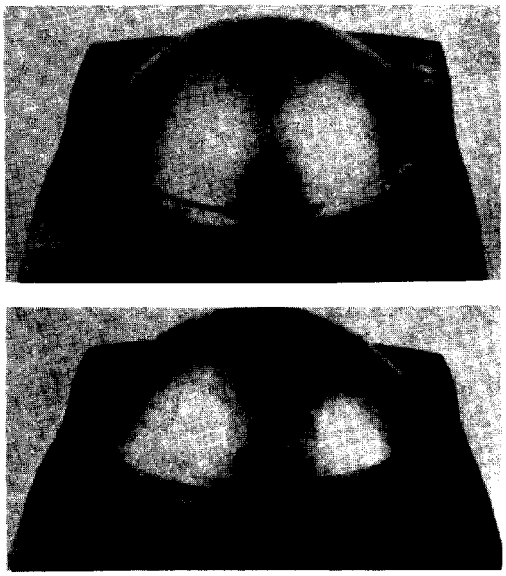


FIG. 2 — Comparison of annealed cast steel plate (top) and annealed ductile iron plate after explosion impact.

Table I—Grades of Ductile Iron

Grade	Minimum Mechanical Properties [1] Tensile Strength psi	Yield Strength psi	Elongation in 2", %	Matrix Structure
60-40-18	60,000	40,000	18	Ferrite
65-45-12	65,000	45,000	12	Ferrite
80-55-06	80,000	55,000	6	Pearlite
100-70-03	100,000	70,000	3	Pearlite
120-90-02	120,000	90,000	2	Fine carbide dispersed in ferrite

[1] *ASTM Specification A536-67*

Table II—Range of Ductile Iron Compositions

Type	Composition, % T.C.	Si	Mn	Cr	Ni	P	S	Mg
Low alloy	3.0-4.5	1.5-4.5	0.2-0.8	—	0.0-2.0	0.02-0.15	0.01-0.03	0.04-0.1
High alloy	2.5-3.0	1.5-6.0	0.4-1.8	0-5.5	18.0-32.0	0.0-0.2	0.01-0.03	0.04-0.1

been classed as weldable by several different welding processes. It is also generally conceded that because of its metallurgically superior quality, ductile iron should be, and is, inherently easier to weld than gray iron. While this may be true, reasonable care is required in the welding of ductile iron.

One of the reasons that gray iron has been welded satisfactorily is that relatively little is demanded of this material in terms of mechanical properties. Most gray irons are brittle and relatively weak and service stresses are mainly compressive. Hence, for such material there is no need for ductile joints or high-strength weld deposits. Because ductile iron offers superior base metal properties, it is often subjected to tensile loads. Weaknesses which would not be important in a gray iron joint are intolerable in a ductile iron joint.

Although metal-arc welds in ductile iron develop excellent tensile strength and the joint properties are adequate for most purposes, the formation of a carbide flash at the base metal-weld metal interface, and martensite in the weld heat-affected zones, has sometimes given rise to machining difficulties[2] even though the weld deposits themselves are easily machined. Additionally, metal-arc weldments in ductile iron have been considered unsuitable for applications involving severe shock or high fluctuating stresses.

Formation of a zone of extreme hardness may be avoided by oxyacetylene welding the casting with ductile iron filler rods, but for development of optimum ductility in gas welds in ferritized ductile iron, full postweld annealing is essential. In this country, a cerium-containing rod[3] is available, while in Europe the filler metal contains magnesium and 5% nickel.[4]

Heat-Affected Zone. Control of the microstructure in the HAZ is one of the most significant of the problems encountered in attempting to obtain welded joints in ductile iron with properties which match those of unwelded material. A number of investigators[4,6,7] have described microstructures which are present in the heat-affected zones. Because of the inherently faster cooling rates, these undesirable hard structures are present to the greatest extent in welds made by the metal-arc process. The microstructures are complex and usually consist of a mixture of martensite, austenite, primary carbide and eutectic, surrounding partly dissolved nodules of graphite. The total width of the hard zone varies somewhat, but is usually of the order of 0.03 to 0.10 in. Hardness in the as-welded condition ranges from 300-700 VHN, but areas with hardness values of above 450 VHN constitute a relatively small percentage of the total.

The mechanism by which these structures are formed is well established. Under the influence of heat from the arc and the molten weld metal, a narrow zone adjacent to the weld metal becomes heated to temperatures just above the eutectic temperature. In reaching this temperature, the matrix is first austenitized and then rapidly begins to dissolve the nodules of graphite. This results in a car-

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Table III—Ferritize-Anneal Heat Treatment

Hold 3 hrs. at 1550-1650°F Furnace cool to 1275°F
Hold 5 hrs. at 1275° Furnace cool, slowly, to 1100°F
Air cool to ambient temperature

Table IV—Transverse Joint Properties-ENiFe Electrode

Type	*Plate Condition*	*Tensile Strength, psi*	*Yield Strength, 0.2% offset, psi*	*Elong, %*	*Reduction Area %*	*Weld Hardness, Bhn*	*Max. Hardness—Heat Affected Zone, VHN*	*Bend Angle, deg.*	*Unnotched Charpy Impact Strength, ft-lb*
Ferritic (60-45-10)	Unwelded	68,500-72,500	47,500-57,000	10-18	—	—	—	78-80	101
	As-welded	51,000-68,500	40,000-57,000	1-9	11-18	205-250	665	14-28	14-20
	Annealed[1]	54,000-72,000	37,000-57,000	5-14	5-28	185-235	420	18-38	28-36
	Ferritize annealed	61,000-68,000	39,000-52,000	6-12	3-5	175-180	175	40-45	33-36
Pearlitic (80-60-03)	Unwelded	98,500	86,000	3	Nil	—	—	5-10	10-11
	As-welded	66,000-82,000	71,000-76,000	1-2	3-4	190-225	535	14-15	13
	Annealed[1]	63,000-71,000	68,000	1-2	2-6	175-205	555	—	11
	Ferritize annealed	57,000-65,000	49,000-51,000	6-12	7-15	175-180	185	33-36	36-40

[1] 1100-1200°F, for various times.

bon-rich austenite and small pools of carbon-rich liquid. Then as the weld metal and the heat-affected zone are rapidly cooled, under the influence of the large mass of cold metal just beyond the peak-temperature zone, the carbon-rich austenite transforms partially or wholly to martensite and the liquid freezes as a mixture of eutectic and primary carbide. Annealing in the range 1,000-1,200°F is sufficient to transform the martensite to softer structures. Still higher temperatures completely ferritize the acicular material.

Effect of Casting Structure. With respect to the maximum hardness attained in the heat-affected zone, it appears to make relatively little difference whether the original matrix structure is ferritic or pearlitic. However, the total width of the heat-affected zone is greater in the case of the pearlitic iron, so that the gross effect is more evident when a pearlitic iron is welded.

The amount of eutectic melting and, therefore the amount of primary carbide which is formed on heating, is probably about the same in either case, but the pearlitic matrix is more readily converted to carbon-rich austenite and thereby becomes a more potent source of martensite. Since it is during this transformation to martensite that underbead cracking is likely to occur, trouble of this type is more probable in the pearlitic grade than in the ferritic grade. This behavior is normal if the matrices of the two irons are thought of as low-carbon ferritic steel and high-carbon pearlitic steel, respectively.

Casting Defects. In evaluating the weldability of any given composition or individual casting, care must be exercised that underbead cracking difficulties are properly accessed. Dross inclusions and spongy or poorly fed areas can be sources of difficulty. If the casting to be welded is metallurgically unsound, the weldment is almost certain to be of poor quality.

Detection of these defects is not always easily accomplished. For critical parts or where pressure tightness is required, careful inspection assumes increased importance. In some instances, an experienced weldor can detect sponginess by the manner in which the base metal "collapses" when exposed to the welding arc. In such cases, it is important to stop welding and grind or chip to sound metal.

Metal-Arc Welding

Nickel-Iron Electrodes. Most of the test work and actual fabrication of ductile iron have utilized nickel-iron electrodes conforming to AWS Specification A5.15-65T (ASTM A398-65T), under the designation ENiFe. This type appears to produce the best results both as to optimum properties and consistency of performance. The range of properties shown in Table IV has been condensed from the published data.

Even with this electrode, which seems to be optimum, there is a disturbing amount of scatter in the reported values for both strength and ductility. The greatest variation is noticeable in comparing the results obtained in different laboratories, but indi-

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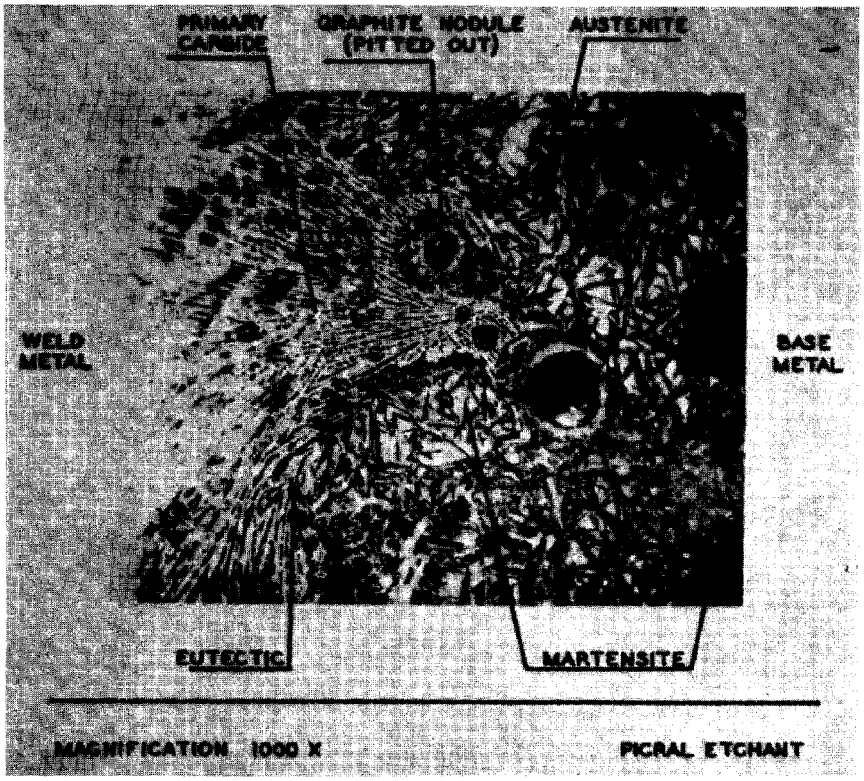
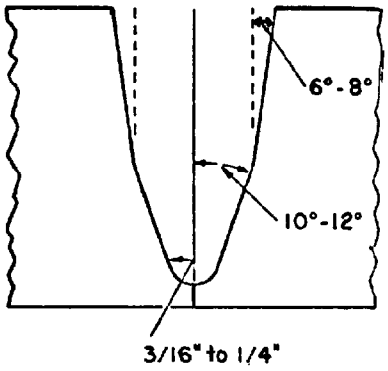


FIG. 3 — Typical microstructure of a weld heat-affected zone in ductile iron.

FIG. 4 — Modified U-joint is utilized for welding heavy sections of ductile iron.



vidual investigators noted nearly as much. In the AWS test program on welding iron castings (1965), the same problem was experienced. While some of the variations may be inherent in the cast or ductile irons, others are definitely due to lack of weldor training with this type of electrode.

Other-Coated Electrodes. The ferritic type of electrode is less expensive than the nickel-iron type, but this advantage must be weighed against its greater crack sensitivity.

Various copper-base electrodes have been tried but with poor results. The major difficulty appears to be an inability to effect a good metallurgical bond between the iron and the weld metal. Fusion-zone cracking and other defects are likely to be present.

Techniques

Preparation of Castings. The casting skin and all adhering foreign matter must be removed from the surfaces to be joined and from the area immediately adjacent. This can be done by machining, grinding or other suitable means. Thorough cleaning will decrease the likelihood of contamination and improve slag coverage and subsequent removal characteristics.

Castings which have been in service are often impregnated with oil or grease. These should be removed by suitable solvents, steam cleaning or similar means. Where possible, heating the casting for 30 minutes at 700°F or 15 minutes at 1000°F will help to eliminate volatile matter and decrease gas evolution during welding.

Castings containing gas should be baked at 1000°F for at least five minutes, if possible. If furnace heating is not practical, the weld area may be locally heated with large acetylene torches. If neither of these methods can be used, a third procedure is useful:

1) Deposit a root pass. This pass will be porous and should be removed by grinding.

2) Lay down another root pass. If this pass is porous, remove it also.

3) Continue this procedure until the heat from welding has driven out all of the gas and a sound bead is obtained.

4) Finish the weld in the usual manner.

Defective areas should be cleaned out down to sound base metal before the repair is attempted.

Joint Design. Joint designs similar to those used for mild steel are suitable. The dimensions are: 60-70 deg. included angle for V-joints, 3/16-¼ in. root radius and 10 to 12 deg. side angle for U-joints. In heavy sections, a modified U-joint reduces the amount of welding required (See Fig. 4).

Electrodes. Nickel-iron welding electrodes can be operated on either a-c or d-c. When used on d-c, reverse polarity is recommended. In general, the welding current should be as low as possible consistent with smooth operation, good bead contour, and the obtaining of a good "wash." Overhead welding will require a reduction of 5 to 15 amps

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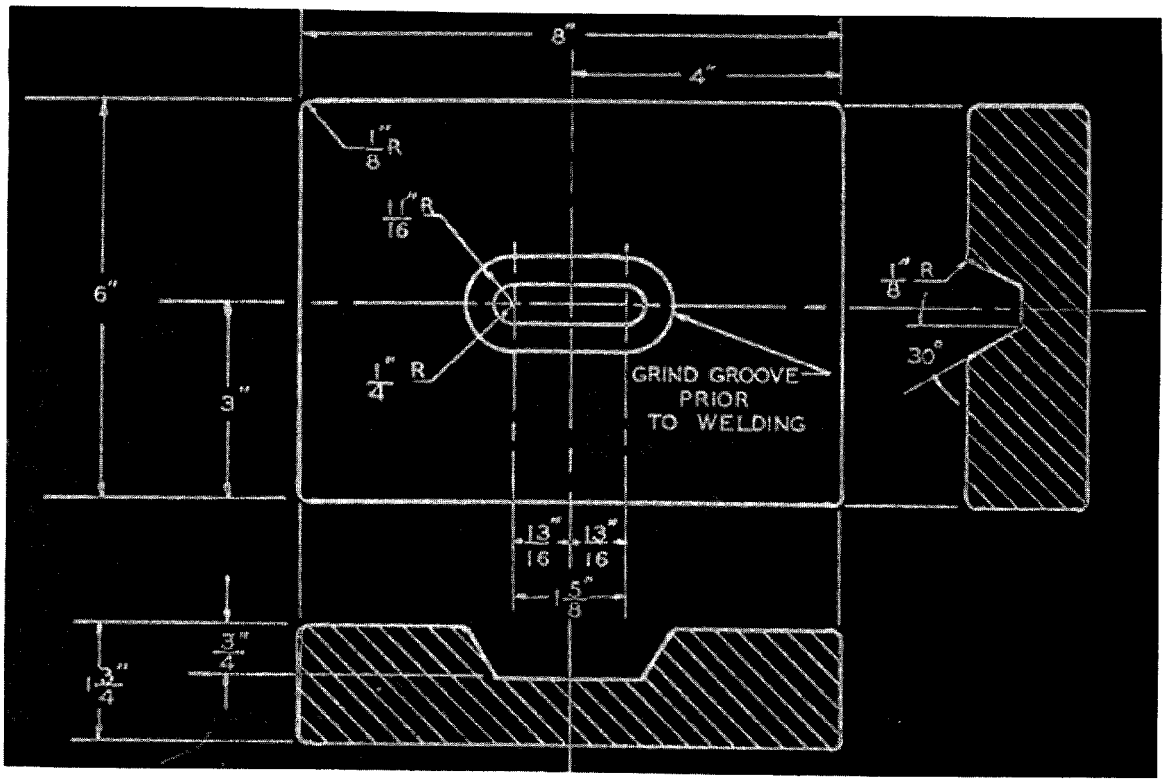


FIG. 5 — Self-restraining test block for evaluating weldability.

and vertical welding, a reduction of 10 to 20 amps. The manufacturers' recommendations for current settings should be followed.

Welds can be deposited in the flat, vertical and overhead positions by either beading or weaving. The weave, when used, should not be more than three times the diameter of the electrode. When welding in grooves, better wash will be obtained on the side of the joint if a weaving technique is used. A skip technique may be used to minimize distortion in extensive repair welds, extra-long welds or heavy sections.

Arc length should be about 1/8 to 3/16 in., the same as used for mild steel welding electrodes. The flux should be removed from each crater before making a restrike and completely removed from each bead before starting a subsequent bead. This is readily done by scratching with the corner of a cold chisel and wire brushing.

These electrodes, like all others, should be kept dry. The special sealed containers in use offer a high degree of protection but they should be stored in a dry place.

An excellent manner of storing electrodes, particularly when the container has been opened, is in a closed cabinet containing desiccants such as indicating-type silica gel. Heating the cabinet with an electric light bulb or a strip heater to a temperature of 15 deg. above room temperature is another effective procedure. Temperatures in excess of 125F are seldom needed.

Damp electrodes can be re-dried by baking. The time and temperature should be as indicated by the electrode manufacturer.

Preheat. Using the test block shown in Fig. 5, Schumbacker and Schaeffler[5] established the need for a minimum preheat of 250-350F for the 60-45-10 grade of iron welded with ENiFe electrodes. A preheat temperature up to 600F may be desirable where unusually high welding stresses are anticipated. In either case, the effect of the preheat is to minimize the possibility of cracking in the heat affected zone or in the weld. Strength and hardness values are relatively unaffected.

Preheats above 600F increased the amount of primary carbide and the depth of the heat-affected zone. Up to that temperature, there was no significant effect on the microstructures.

Postheat. Postweld anneals have very little effect on either the hardness, strength or ductility of the deposited weld metal. A ferritize-anneal is necessary, after welding, to decompose the carbide structures in the heat-affected zone. During this anneal the Fe_3C decomposes and the carbon is rejected as extremely small graphite spheroids, frequently arrayed in a chainlike pattern which impairs the ductility. In a fully ferritized joint where, for instance, the elongation of the base metal might be of the order of 15-20%, the elongation values across a welded joint will seldom exceed 5-10%. The impact toughness and bend ductility are likewise not fully restored. If necessary, a 1,275F annealing treatment or a low tempering treatment (1,000F) can improve the machinability of the weld-base metal interface. ●

The second part of this article deals with oxyacetylene, short-arc and dissimilar metal welding.

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Welding ductile iron—Part Two

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IN PART ONE of our discussion on welding ductile iron (WE Nov., page 57) we focused on the metal's general properties and what affect welding had on these. One welding process in particular, manual metal-arc welding, was given a close look.

In this second and concluding part we will look into the welding of ductile iron by the oxyacetylene method and welding by short-circuiting arc. A discussion on joining ductile iron with dissimilar metals using coated electrodes and short-circuiting arc methods is also included.

Oxyacetylene Welding

Ductile Iron Filler Rod. Both magnesium-bearing and cerium-bearing ductile iron filler rods are capable of depositing weld metal with a spheroidal graphite structure. The preference seems to be for the cerium-bearing type which appears to be less prone to porosity than the magnesium-bearing filler rod. Recently, a magnesium-bearing filler rod containing 5% nickel has been developed, in which the porosity difficulties have been eliminated. Deposits made with this nickel-containing rod are less prone to contain primary carbide than are nickel-free deposits.

Smaller amounts of the hard constituents are formed in the heat-affected zones of oxyacetylene welds than in the heat-affected zones of metal-arc welds. Although the peak temperatures may be of the same order of magnitude, the cooling rate is very substantially less in the oxyacetylene process. Nevertheless, sufficient primary carbide and acicular structures are likely to be present to create hardness levels in excess of 320 VHN in both the heat-affected zone and the weld deposit. The usual practice is to retard the cooling rate of the welded structure by suitable insulation or by other means. However, an anneal at 1600F is usually necessary to ferritize the structure completely.

Table V shows a summary of the properties of oxyacetylene welds. Like those obtained with ENiFe electrodes (Table IV), these tests results are characterized by considerable variation in strength and ductility values, and in no case is a level of joint ductility obtained which matches that of the base metal.

Because of the lesser amount of hard structure formed, the oxyacetylene process may be preferred over metal-arc welding in some instances. Casting defects in raw castings, for instance, are sometimes corrected in this manner. In some instances, where the massive heat pattern of the oxyacetylene process and the resulting distortion could be troublesome, as in finished or semi-finished parts, the metal-arc process is preferred in spite of the somewhat higher hardness levels in the heat-affected zone.

Copper-Base Filler Rod. A few investigators have considered the possibility of braze welding with copper-base alloys using the oxyacetylene process, but contrasted with some rather good experience with tightly fitting brazed joints, the results obtained by braze welding have been consistently poor. As in the case of metal-arc welds made with copper-base covered electrodes, the joint is likely to be weak at the line of fusion.

Short Circuiting Arc Welding

Short circuiting arc, a comparatively recent variation of Mig welding has been tried on ductile iron[8]. This process requires a lower energy input and develops a smaller weld pool than regular Mig. It was considered that the machinability and toughness of arc welds would be improved if the hard zone could be made discontinuous. Although the width of the hard zone is diminished by controlling the heat input during the metal-arc welding with flux-coated electrodes, discontinuous hard zones are seldom found in the heat-affected zones produced

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Table V—Properties of Oxyacetylene Welds-Ductile Iron Filler Rod

(Annealed, Ferritic D.I. Plates)

Postweld Heat Treatment	*Tensile Strength, psi*	*Yield Strength, 0.2% offset, psi*	*Elong. in 2-in. %*	*Weld Hardness, Bhn*	*Max. Hardness—Heat Affected Zone, Bhn*	*Unnotched Charpy Impact Strength, ft-lb*	*Location of Fracture*
Transverse joints							
Annealed (1000°F)[1]	51,000-56,000	—	1	210	305	—	Weld & fusion zone
Annealed (1650°F)[2]	41,500-69,000	45,000-53,000	1-6	—	—	—	Weld & base metal
Full ferritize annealed	49,000-64,000	—	—	—	—	—	—
All-weld-metal							
As-welded	57,000	54,300	3.5	200		77	—
Annealed (1100°F)[3]	76,900	60,700	6.8	—		79	—
Annealed (1600°F)[3]	66,200	47,400	12.5	150		67	—

[1] 1000°F/2 hr — furnace cooled
[2] 1650°F/4 hr — furnace cooled
[3] Time unspecified

in commercial ductile iron.

It has been established that short-circuiting arc welding of ductile iron with certain low-carbon nickel-alloy filler wires markedly limits the width of the hard zone. In weldments made by the process, the carbide phase is frequently restricted to relatively thin envelopes surrounding graphite spheroids, and the properties of the joints are improved.

Butt Welds. Butt welds were made in ductile iron plates, ½ and 1-in. thick, using 0.030-in. diameter nickel alloy filler wires. Of these filler wires, the Nickel Filler Metal 61* (AWS A5.14, Class RN41) and Monel Filler Metal 60* (AWS A5.14, Class ER60) are grades normally employed for welding the respective nickel alloys by the argon-shielded processes. Some of the ductile iron plates were welded in the "as-cast" condition; others were ferritized (2 hrs. 1650°F + 4 hrs. 1275°F, air cooled). If the interpass temperature exceeded 250°F, the plates were allowed to cool to this temperature before welding continued. Radiographic examination of these welds demonstrated the high quality obtained with this process.

Welds made using filler metals of mild steel or high nickel alloys and argon-20% CO_2 as the shielding gas were unsatisfactory. Severe cracking occurred in the mild steel weld deposits. The nickel alloy weld metals contained numerous pores and suffered from gross oxidation.

Microscopic examination of sections cut transverse to the 1-in.-thick butt welds showed that the carbide zone at the weld-fusion line was discontinuous (Fig. 6) and was, in general, confined to envelopes around the graphite spheroids nearest to the fusion line. Tempered martensite was also observed in the pearlitic ductile iron plates near the weld-fusion lines, but in the ferritized irons it occurred mainly near the graphite spheroids. Un-

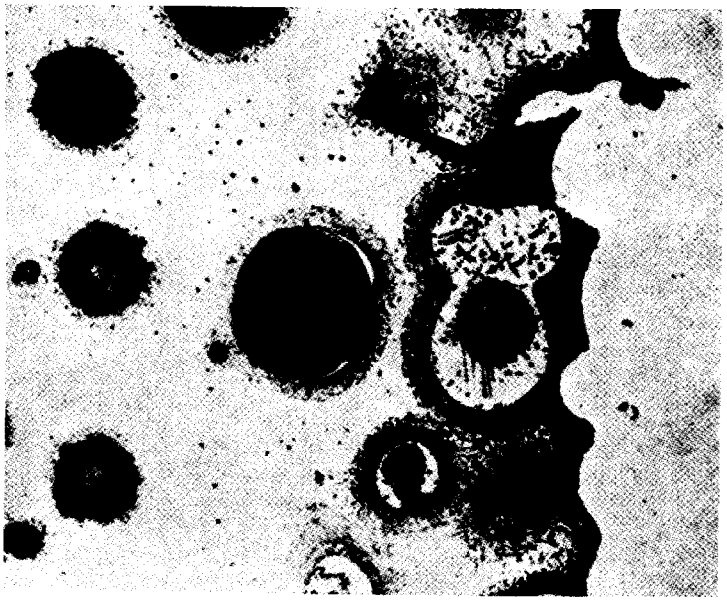


FIG. 6 — Short-circuiting arc weld in ductile iron. Note restriction of carbide phase to thin envelopes surrounding the graphite spheroids.

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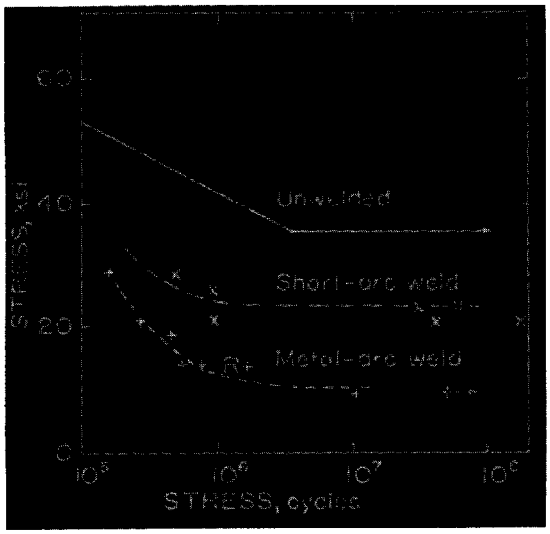


FIG. 7 — Fatigue strength of plain and welded ferritized ductile iron.

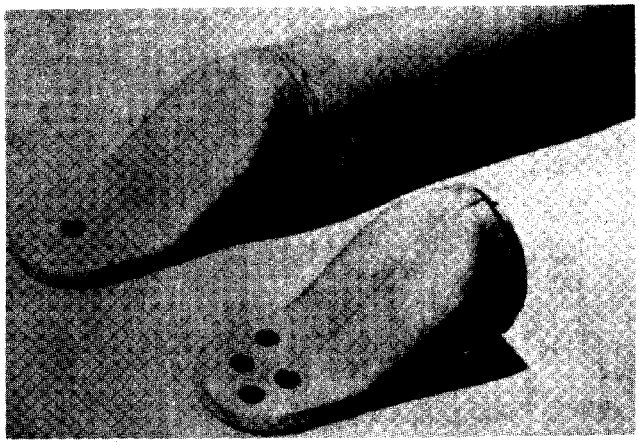


FIG. 8 — Ductile iron clevis cap welded to carbon steel tube with ENiFe electrode (photo approximately one-third actual size).

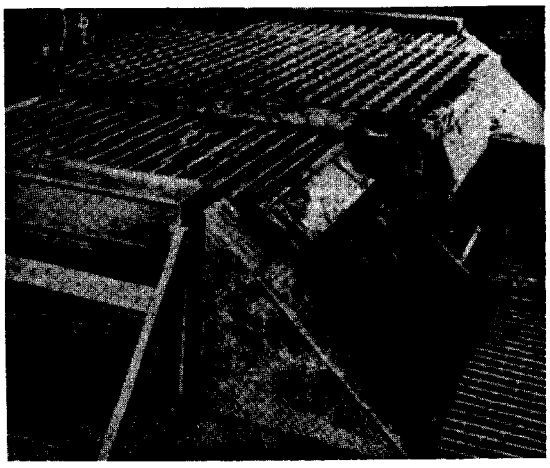


FIG. 9 — Foundry shakeout deck required welds of ductile iron to itself and to steel side decks.

tempered martensite was detected next to the capping runs at the surfaces of the joint.

An interesting feature of these trials was the production of crack-free welds in the pearlitic ductile iron without the use of preheat. Metal-arc welding of the pearlitic iron usually requires preheat temperatures above 400F. Also, the flux-coated electrode may require baking at 500F to prevent the possibility of delayed cracking, due to hydrogen, in the pearlite matrix.

The transverse fatigue strengths of test pieces machined from short-arc butt welds in ferritic ductile iron made with the nickel filler metal in 1¼-in.-thick ferritized ductile iron plate are superior to those of specimens produced by the metal-arc process with a commercial nickel-iron flux-coated electrode (Fig. 7). Joints made in normalized ductile iron by the two processes tended to show similar fatigue strengths.

Pipe Welding and Other Aspects. Ferritized centrifugally cast ductile iron pipe is in substantial and rapidly expanding use in construction of pressurized pipe lines, and the suitability of short circuiting arc welding has been explored in that context. For positional welding, the deposition characteristics of the Monel Filler Metal 60 proved to be better than those of Nickel Filler Metal 61.

Argon-shielded short circuiting arc welding of a pipe 4-in. diameter x ¼-in. wall, with the pipe axis in the horizontal fixed position was successfully performed. Because of the low penetrating power of the short circuiting arc, it was found important to maintain the root gap during deposition of the root run. The joints were therefore tacked before welding. The runs were deposited by the downward vertical technique, from the 12 o'clock position. Satisfactory butt joints were also produced in the pipe rotated under the torch with both filler metals. The carbide at the weld-fusion line occurred in somewhat greater amounts than in welds made by the short circuiting arc process on ½-in. and 1-in.-thick ductile iron plates. This increase was probably caused by the greater build-up of welding heat in the comparatively thin-walled pipe.

Dissimilar Metal Joints

Coated Electrode. When welding ductile iron to a dissimilar metal, the best results have been obtained using a covered electrode of the ENiFe type. Strong crack-free joints have been prepared between ductile iron and a number of other alloys including mild steel, stainless steel, Monel* (nickel-copper alloys), nickel, and Inconel* (nickel-chromium-iron alloys), with this electrode. The hard zone persists, of course, in the ductile iron side of the joint, but the overall joint ductility is improved slightly because the hard structures are mainly limited to the one side. The weld metal and HAZ of the other material generally remain soft and can yield to relieve stresses.

Of the several possible dissimilar metal joints,

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the one between ductile iron and mild steel is the one most likely to be encountered. Results have been satisfactory with the ENiFe electrode. Examples are shown in Figures 8 and 9.

Short Circuiting Arc. The short circuiting arc technique offers a useful process for dissimilar metal combinations involving ductile iron. Values in Table VI demonstrate that a short circuiting arc butt weld made between pearlitic ductile iron and a mild steel plate developed satisfactory transverse tensile properties. As shown in Figure 10, a fillet weld between these materials (mild steel vertical members) did not fail under impact. In Figure 11, an automatic setup for welding a mild steel steering column to a ductile iron steeding gearbox is shown. The welding process is fully automatic short circuiting arc using Nickel Filler Metal 61.

Table VI—Transverse Tensile Properties-Short Circuiting Arc Welding Mild Steel-Ductile Iron

Filler Metal	*U.T.S. ksi*	*0.2% Y.S. ksi*	*Elongation, % 2 in.*	*½ in*
Nickel	71.2	43.0	15.5	4.0
	71.2	45.6	13.0	4.0
	72.5	44.3	12.0	4.0

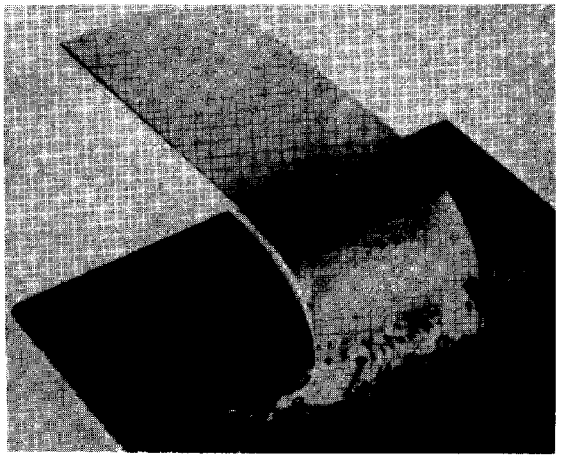


FIG. 10 — Short-circuiting arc fillet weld between mild steel and ductile iron (Nickel filler metal 61).



FIG. 11 — Welding a mild steel steering column to a ductile iron steering gearbox using fully automatic short-circuiting arc welding process and Nickel filler metal 61.

Summary

While ductile iron can be joined by most of the commercially accepted welding processes, the major portion of welding in industry has been with ENiFe coated electrode. The next most successful process has been oxyacetylene welding using a ductile iron filler metal. The main problem when welding ductile iron is the formation of primary carbides and martensite in the weld heat-affected zone. While preheating or postweld annealing will reduce the deleterious effect of martensite, only a ferritize-anneal heat treatment of the weldment can eliminate the carbide structure. While the presence of such hard constituents is not desirable, weldments with this condition have given satisfactory service in many applications. Also, when ductile iron is welded to steel, greater ductility is achieved because the hard zone is present on only one side of the joint and stresses are relieved by yielding of heat-affected zone in the mild steel. It is most important to consider the properties of weldments in relation to the service requirements of the part.

The latest development in the field of ductile iron welding is the use of the short circuiting arc welding process. When this process, which involves a lower heat input, is used, the carbide phase is frequently restricted to relatively thin envelopes surrounding graphite spheroids and the mechanical properties and machinability of the joints are improved. It appears that the plate thickness has some effect, fewer carbides forming in heavier plates. ●

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2. Kihlgren, T. E., and Waugh, H C., "Joining of Ductile Iron by Several Welding Methods", The Welding Journal, 32 (10) 947-956 (1953).
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8 Osman, J. L., Stephenson, N., and Thorneycroft, D. R., "Development in the Welding of S. G. Iron", The Ductile Iron Bulletin, No. 11 (Nov. 1965).

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AWS WCI 85 ■ 0784265 0006978 4 ■

Repair welding of ductile iron

By: A. J. Osborn
Senior Applications Engineer
Eutectic & Castion
Eutectic Corporation

What are Ductile Irons?

Today's trend for castings is that they provide a combination of mechanical and physical properties between those offered for grey-type cast irons and cast or wrought steels. A material bridging the properties of grey irons and cast steels is today known as nodular iron, nodular graphite iron, spheroidal graphite iron, to which the generic term ductile is applied.

The increasing tonnage of ductile or nodular cast irons now being produced, and its wide usage for a broad range of industrial applications, attests to the importance and acceptance by industry of this iron as a useful material.

Ductile Irons

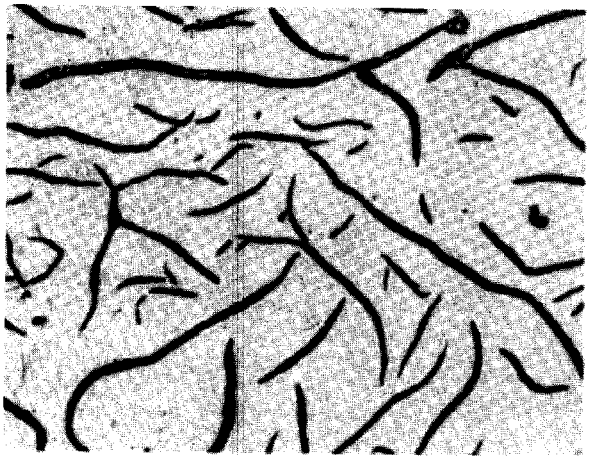
The terms "ductile iron," spheroidal graphite iron," and "nodular iron" all refer essentially to the same engineering material. In these situations where the term "ductile" is used, the reference is to the specific mechanical property of ductility measured as a specific amount of elongation. In many cases, ductile irons have mechanical properties approaching that of low carbon and plain carbon steels. Tensile strength values can range between a low of 340 MPa to a high of 1030 MPa in the as-cast condition, with yield strengths varying between 275 MP and 820 MPa. Ductility values, as may be expected, decline with an increase in tensile strength. The as-cast condition produces higher strength cast iron and lower ductility than do parts that have been heat-treated after casting.

A micro-structural comparison between an ordinary grey cast iron and ductile iron shows that graphite in grey irons is in the form of flakes and in ductile iron, the graphite takes the form of tiny balls or spherulites of characteristic structure.

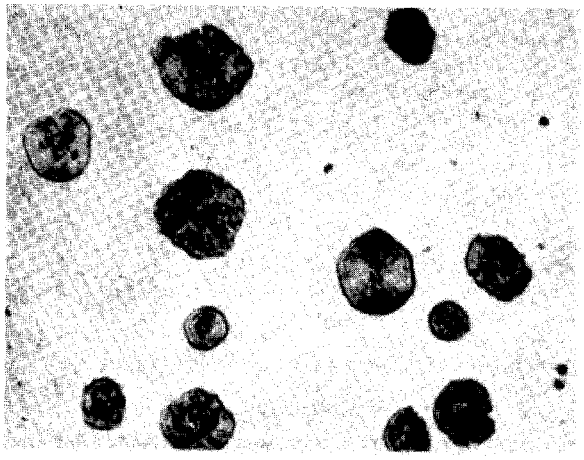
"Grey irons" cover a series of euteciferrous alloys offering a fairly broad selection of mechanical properties with the matrix structure generally being either pearlitic or pearlitic-ferritic. The objection to their use for highly stressed components is that the morphology of the graphite—particularly when it is interconnecting, acts as stress raisers, reducing as a consequence the material's ability to tolerate stresses much above a nominal 300 MPa.

Source: Casting Engineering, Fall 1977

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Microstructure of grey iron showing graphite flakes (Mag. X 100).



Microstructure of ductile iron showing graphite spherulites (Mag. X 100).

Table I — General Analyses of Grey Iron and Ductile Irons

IRON TYPE	C	Si	Mn	P	S	Ni	Cu	Other
	%	%	%	%	%	%	%	%
1) Grey	3.0 -3.35	2.0 -2.40	0.6 -0.7	0.2 max.	0.2 max.	—	—	—
2) Nodular	3.20-4.10	1.8 -2.80	up to 0.80	0.10 max.	0.03 max.	—	—	See "2" Below

2) Additionally, cerium in amount of a few/1000 of 1% or magnesium not in excess of 0.80%

Types of Ductile Irons

Ductile irons are available today offering a wide range of matrix structures and properties achieved either through alloying and/or alloying with heat treatment. To give some idea of the change in mechanical properties effected through alloying and heat treatment. Table II highlights some major properties:

Table II — (N.B. Tensile & Yield Strength Values)

(N.B. Tensile and yield strength values are given in Megapascals)

		MECHANICAL PROPERTIES					
IRON CLASS	DESCRIPTION	DUCTILE IRONS AS CAST			FLAKE-GRAPHITE AS CAST		
		UTS	YP	ELON.	UTS	YP	ELON.
Pearlitic Iron	Structure is mixed ferrite and cementite with carbon spherulites	690	540	1-5%	340	—	—
Ferritic Iron	Structure is mixed pearlite and ferrite with the carbon finely dispersed	540	380	12-25	340	—	—
Alloyed Irons							
Accicular Iron	Generally alloyed with nickel/ molybdenum/ chromium to produce accicular, austenitic or martensitic structures	980	780	3%	340	—	—
Austenitic Iron	"	470	280	10-25	340	—	—

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Repair Welding of Ductile Irons

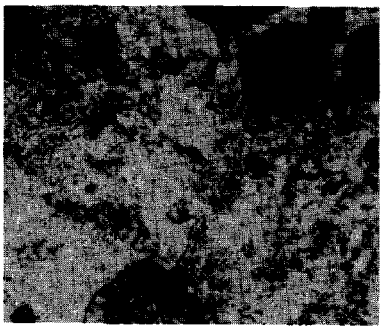
Much has been said about the poor weldability of cast irons compared to plain carbon steel welding when using the Shielded Metal Arc Welding Process. The principal reasons given are that the factors influencing the success, or failure for that matter, in obtaining satisfactory welds, such as interfacial carbide formation, heat affected zone cracking, the low strain-to-failure at temperatures below which plastic deformation occurs, and the large shrinkage stresses involved in fusion welding, prevent the achievement of other than randomly successful repair welds. This is now no longer the case. These problems are now overcome by the proper selection of repair electrode chemistry, control of heat input and a judicious use of proven repair welding techniques.

Selecting a Repair Electrode

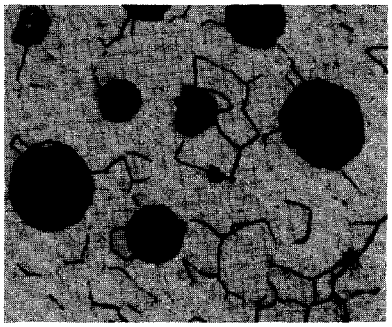
Shielded Metal Arc Welding is, and will probably remain, the most commonly used repair welding process. The lower heat input and the excellent mechanical properties achieved, make it an ideal tool for the repair welding of a wide range of both grey and ductile iron components. However, not all electrodes can be used and the need to know which are suitable and acceptable is mandatory if successful welds are to be achieved.

Shielded Metal Arc Welding Using Ferritic Electrodes

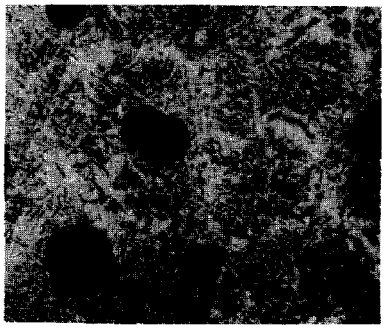
The use of ferritic electrodes for welding ductile iron and even grey irons for that matter, is frought with difficulty. Cracking or extremely hard deposits, single pass-deposits have been measured as high as 500 DPN, makes the use of this chemistry very risky. Residual stresses measured on bead-on-plate with mild steel have been measured as high as 310 MPa in the longitudinal direction and as high as 110 MPa in the transverse direction. It is, therefore, most unlikely that crack-free welds and welds having acceptable machinability can be obtained with this class of electrode unless very high pre- and interpass temperatures are used.



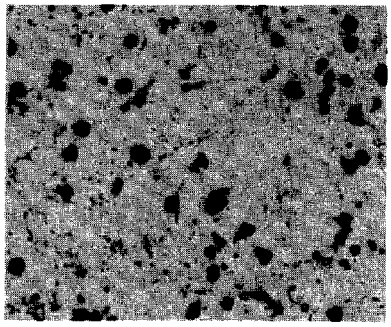
Pearlite Ductile Iron
(Mag. X 250)



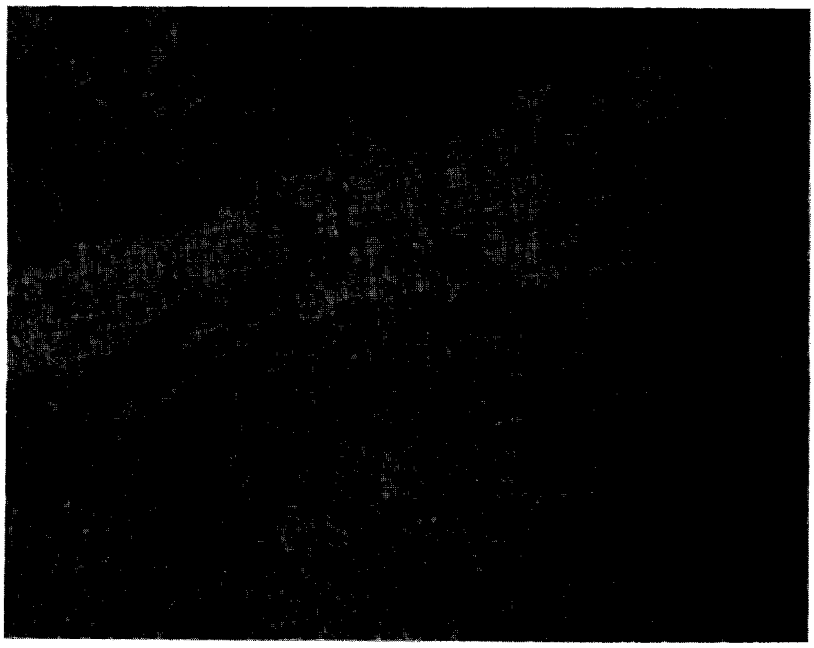
Ferritic Ductile Iron
(Mag. X 250)



Accicular (Ni/Mo) Ductile Iron
(Mag. X 200)



Austenitic Ductile Iron
(Mag. X 100)

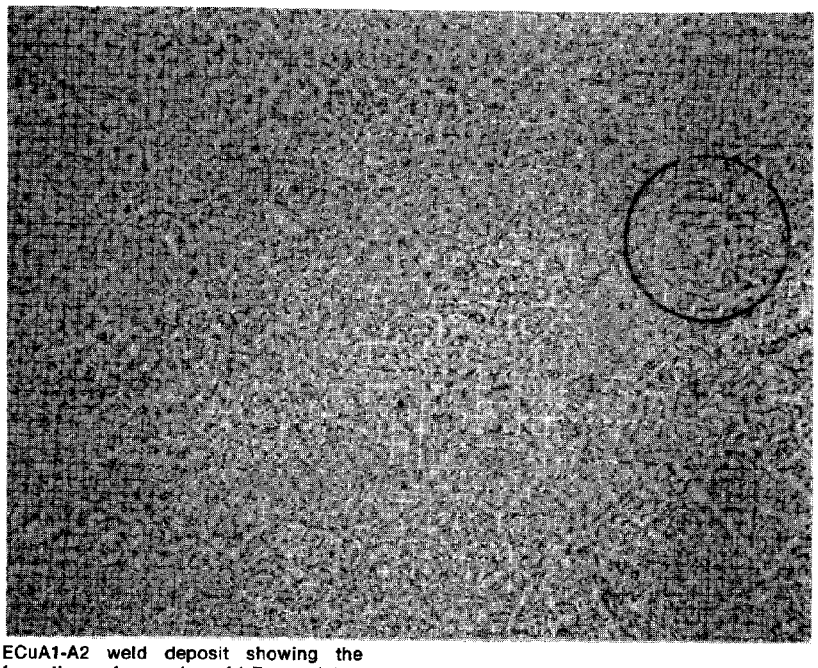


Ferritic weld deposit on grey cast iron. (Mag. X 100)

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Shielded Metal Arc Welding Using Aluminum-Bronze Electrodes (ECuAL-A2)

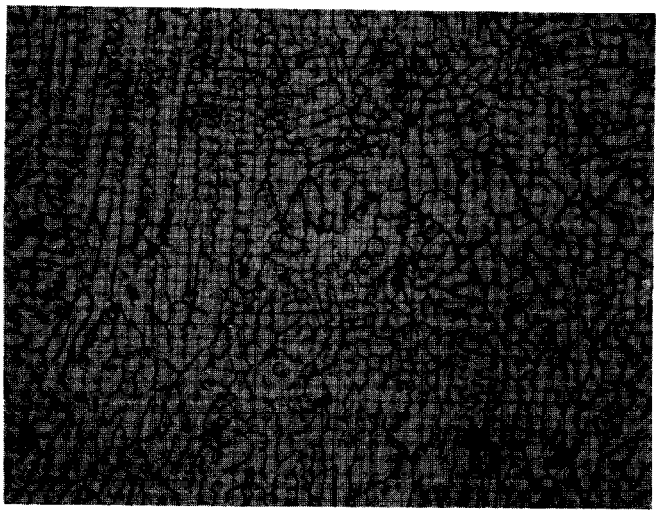
Although aluminum-bronze electrodes have been used successfully to weld ductile irons, their primary use has been for cladding and not for joining. The technique called for with this class of electrodes such as specially prepared joints; preheat and slow-cooling requirements and a rather fast deposition rate, make it a rather difficult electrode to use. The need to obtain adequate fusion and the consequent development of large-size A1:C:Fe nodules, markedly affect elongation values. Aluminum-bronze deposits in the as-deposited condition have given measured elongation values less than 5% making them a rather poor choice for the high strength joining of ductile irons. However, without being too negative toward this class of electrode, it should be borne in mind that they make an ideal cladding surface for such components as slipper shoes, bearing seats in cast iron plummer blocks and impeller edge surfacing.



ECuA1-A2 weld deposit showing the formation of massive A1:Fe nodules. (Mag. X 10)

Shielded Metal Arc Welding Using Austenitic Stainless Electrodes (300 Cr/Ni Series)

Austenitic stainless steel electrodes of types 308, 309 and 312 have been used for repair welding both the pearlitic and ferritic ductile irons where interfacial bonding difficulties have arisen. However, under conditions of severe restraint, heat affected zone cracking is a very pronounced effect. This is thought to occur due to the precipitation of carbon at the ferrite-austenite phase boundaries. Additionally, the high yield point of stainless type electrodes, 312 S/S has a yield point of around 550 MPa, which is a higher yield point than both the pearlitic and ferritic ductile irons, compounds the tendency for HAZ cracking. As a final statement, one should mention that the introduction of chromium from the welding electrode further increases cracking tendencies by increasing the austenitic transformation products and the carbides $Fe_3C \blacklozenge Fe/Cr_3C$.



Deposits on cast iron using a 312 type stainless steel electrode. (Mag. X 100)

Shielded Metal Arc Welding Using High Nickel and Nickel-Iron Electrodes

Electrodes classified under the AWS Specification A5.15-69, sub-group nickel-base electrodes, particularly

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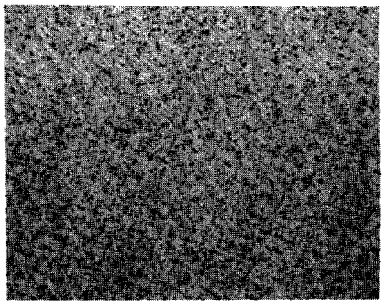
groups ENi-CI and ENiFe-CI, are well known to produce satisfactory mechanical properties when used to weld grey, ductile, nodular and malleable irons. The nickel content in these electrodes provides the strength and ductility with the carbon necessary for graphite rejection being obtained from the electrode coating. Because of the low solid solubility of carbon in nickel, excess carbon is precipitated so that weld deposit microstructures can generally be described as cored, solid solutions containing graphite nodules and aggregates. Although a microstructure containing free graphite aggregates has certain weaknesses under tension, it is possible to obtain a graphite spheroid plus carbide structure through electrode coating modifications. For example, tests carried out on pearlitic and ferritic ductile irons using a modified flux coating nickel iron electrode gave the following average values shown in Table III.

Table III

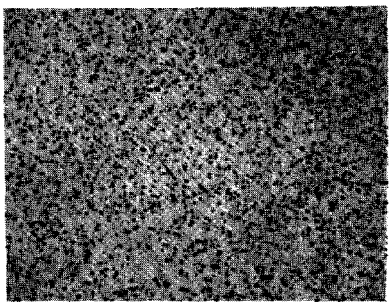
IRON CLASS	ELECTRODE TYPE[2]	MECHANICAL PROPERTIES		
		UTS	YP	ELON.
Pearlitic	XYRON 2-23 (NiFe Type)	550	390	2%
Ferritic	NodulTec 2250 (NiFe Mod.)	450	210	6%

(N.B. Values are given in Megapascals.)

[2] XYRON 2-23 and NodulTec 2250 are patented cast iron electrodes of the Eutectic Corporation.



A nickel-iron weld deposit on cast iron. (Mag. X 100)



A high-nickel weld deposit on cast iron. (Mag. X 100)

The Influence of Heat Input

In any arc-welding process the structural changes occurring in the HAZ are extremely important. For example, Figure XII shows a peak hardness of 420 HV (Rc 40 circa) when a 3-pass weld deposit of a nickel-iron type electrode was used on a pearlitic ductile iron without preheat. On the other hand, preheating to the order of 300°C reduced the peak hardness to a much more acceptable 270 HV (Rc 25). However, it must be stated that a great deal of myth and suspicion surrounds the interpretation of HAZ structures, particularly when carbides are present. The interfacial carbide band, whether continuous or discontinuous, can be reduced by the type of flux formulation used on nickel/nickel-iron electrodes. Continuous carbide bands and such hard transformation products as martensite and bainite are related primarily to the size of the electrode, amperage level, and amount of preheat used. By using preheating techniques judiciously and keeping dilution to a minimum, carbides can be kept discontinuous and undesirable structural changes minimized.

Some Case Histories

Any theoretical discussion of a subject which has essentially a practical leaning should show some application to field problems. The following case histories were developed in Australia; however, they would apply to any number of applications where welding is used as a maintenance tool.

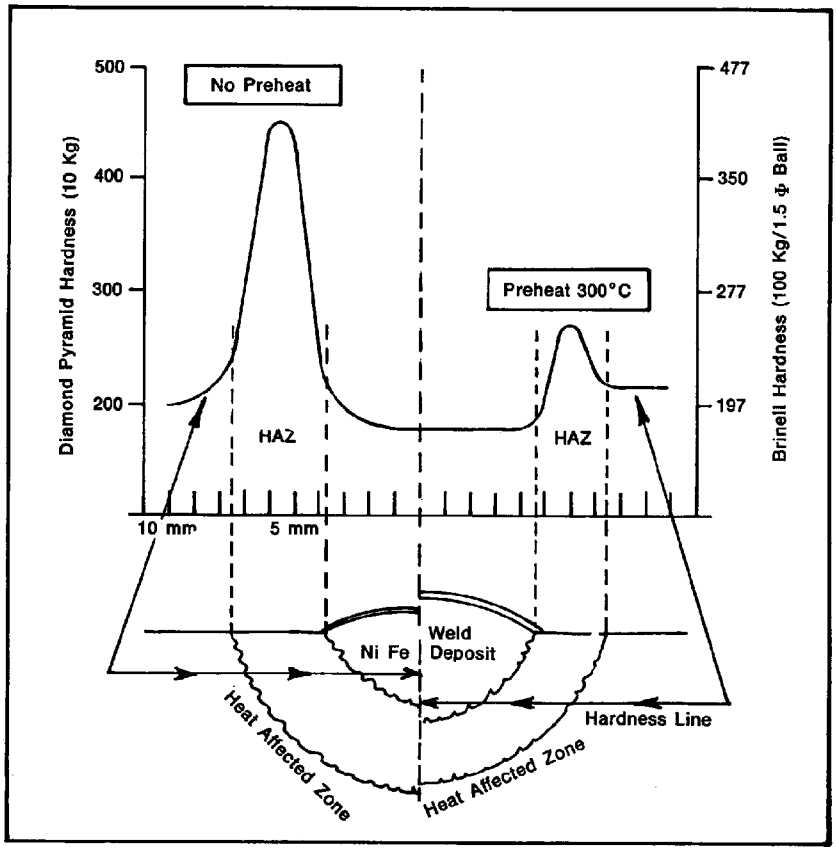


Figure 12
Graph shows the range of hardness for a sample of pearlitic ductile iron welded with and without preheat.

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Case History No. 1

Part Name:
Spur Gear Wheel
Unit Machine:
Drum Winch
Material:
Ferritic Ductile Iron
Reason for Failure:
Maintenance fitter dropped a hammer between the spur gear and pinion drive which resulted in a high stress peak load, sufficiently high to cause a crack to open in two of the spokes and between the two spoked segments.
Preparation:
The gear was first degreased and then dye-penetrant tested for ascertaining the true location and range of the cracks — particularly important with stress rupture cracking. Shop identification after dye-penetrant testing was to centre-punch the crack so that the welder would not miss the crack during chamfering.

Preparation of a crack for welding can be achieved with arc-air, chip-gouging with a pneumatic chisel, grinding or with specially designed gouging electrodes. For this repair, gouging electrodes were used, because of the ease and rapidity with which they remove unwanted metal.

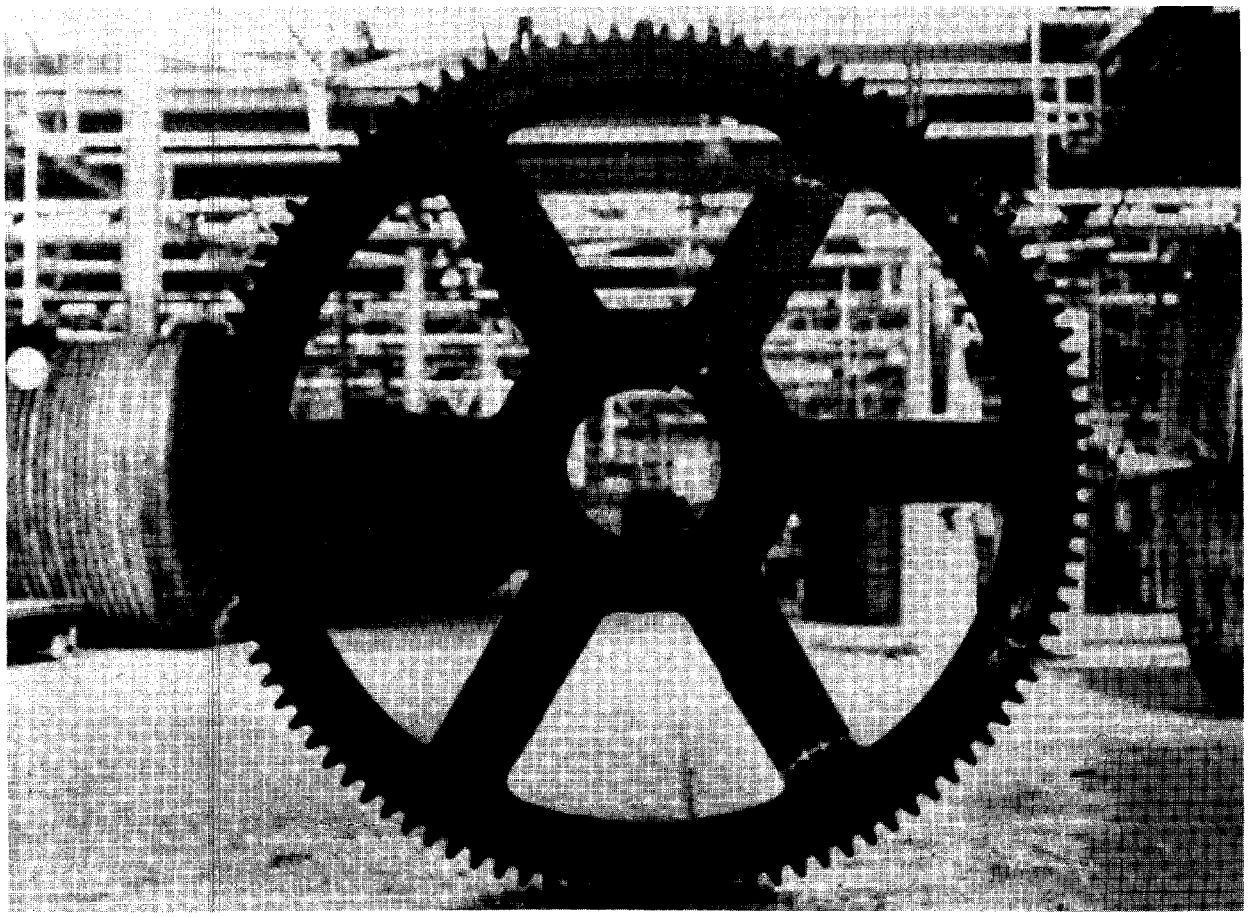
Any process which develops heat during a metal removing operation where the quench effect is severe, can develop martensite/austenite interfacial zones. The removal by grinding of this hard, crack sensitive structure is mandatory, otherwise interfacial cracking can occur.

Although normal practice dictates the use of general preheating — particularly for the non-cracked contiguous spokes, it was decided to preheat preferentially in the immediate area of repair instead of resorting to a total weldment preheat. Furthermore, the fair ductility properties and the sectional restraint were considered advantageous in reducing any tendency for the weld deposit to crack.

Briefly, the welding technique and procedure used to effect the repair was . . .

1. Degrease
2. Dye-penetrant test
3. Chamfer cracks to a single "U" preparation
4. Grind hardened kerf zone approximately 1 mm
5. Preheat locally to 250°C-300°C
6. Deposit weld metal using an ENiFe-CI classification electrode and a back-step technique. Deposit lengths not to exceed a practically determined length of 300 mm. Toe cracking occurred a few times, overcome in this instance by reducing the weld deposit length to 20 mm.
7. Peen all weld deposits other than root and final crown deposits.

Large ductile iron gear wheel repaired with a nickel iron electrode. Diameter — 3 metres.



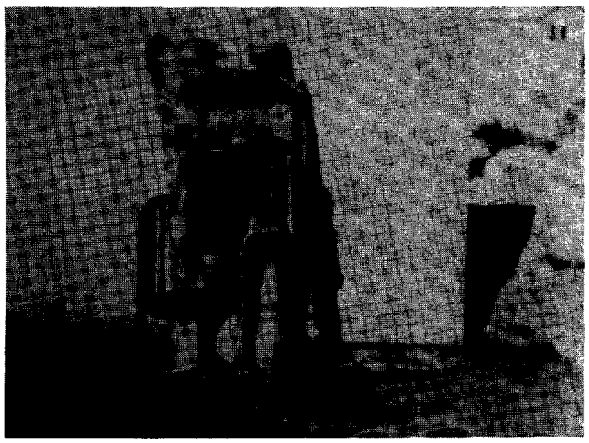
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Case History No. 2

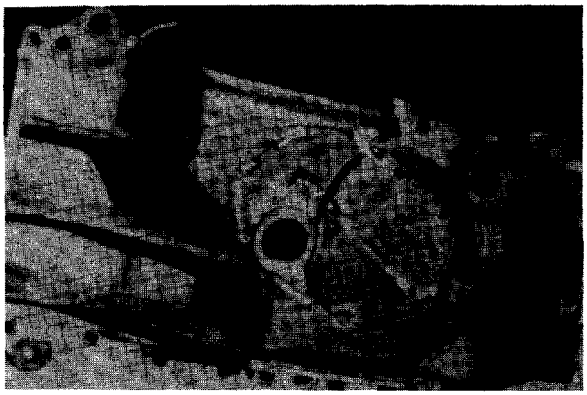
Part Name:
Sump
Unit Machine:
Caterpillar D4
Material:
Grey Cast Iron
Reason for Failure:
During the course of normal operation—always arduous for "cats," the sump crashed into a hard rock outcropping, breaking a rather large section out of the sump bottom.
Preparation:
Cut from 9 mm thick mild steel plate an inseft piece conforming to the general shape of the rupture. Bisect the insert plate as shown in the second continuation photograph. (The technique of cutting in two an insert is simply to make allowance for expansion and contraction when welding the plate periphery to the cast iron sump.)

Back-step welding is recommended for similar repairs as this reduces crater cracking tendencies particularly with such alloys as those meeting the ENNi-CI classification.

The final step after completing the peripheral welds is to join the bisected plates. This can be done with any general purpose mild steel joining electrode such as an E6012, E6013, etc.



The Break!



Insert Welding Technique.



The Completed Repair.

Caterpillar D4 cast iron sump. Welded with a high nickel electrode.

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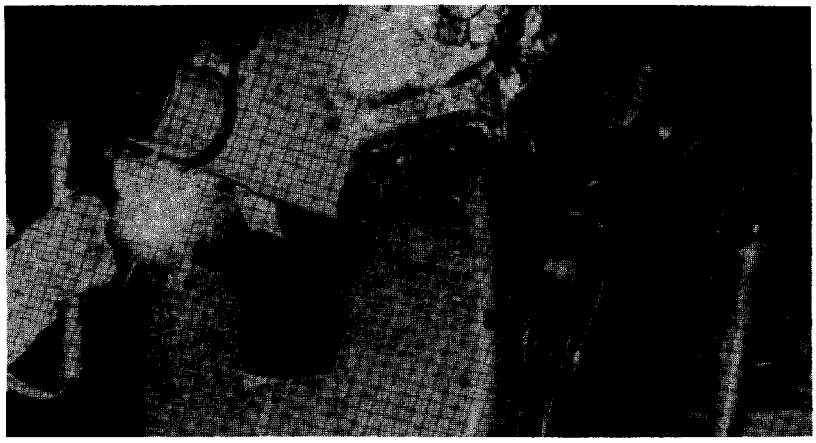
Case History No. 3
Part Name:
Press Frame
Unit Machine:
John Heine Reciprocating Ram Press
Material:
Ferritic Ductile Iron
Reason for Failure:
The classic service error! Trying to blankout from material too thick for the machine size. This resulted in the crankshaft exerting an upward motion of sufficient magnitude to completely fracture one of the side yokes.
Preparation:
In any repair to a machine unit, it is absolutely mandatory that the relative relationship between mating components be checked for an acceptable level of tolerance. This was done by using a spare crankshaft (the original one was bent during the accident) for measuring the respective reciprocating/rotational movements.

Because the break was "clean" i.e. the fracture surfaces abutted without excessive misalignment, the preparation of the joint became a simple matter. A double "U" configuration was obtained using a 6 mm diameter gouging electrode with each joint face ground to remove any HAZ hardened material. To aid in alignment, a small landing of some 2 millimeters wide was left on the top and bottom of each root of the yoke.

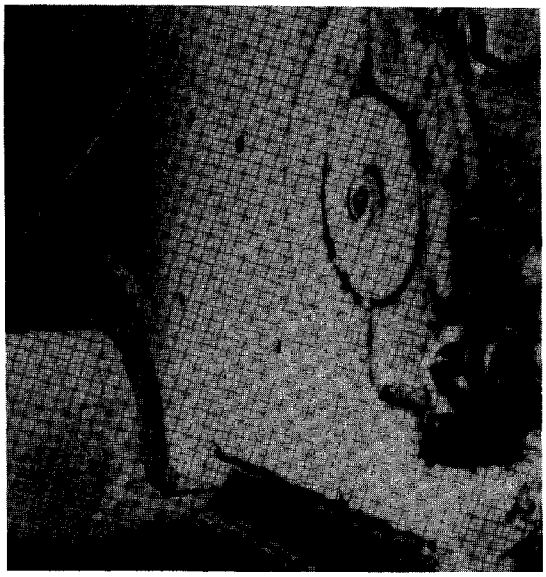
The nature of the break and the massive sectional restraint was thought best to be met by the ENi-Fe-CI classification electrode. Subsequent restraint tests proved the adequacy of this electrode classification for the job.

A "balanced" deposition technique was used and qualified by using flat edge rulers and centre-punched reference points so that monitoring in terms of fixed reference points was readily observed.

The quality of the weld-deposit is readily seen by the close-up photograph.



Repair Malleys side frame, John Heine Press, partially completed.



Close-up repair — Malleys — Sydney.



Shows welder repairing side frame — Malleys, Sydney.

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Welding Nodular Cast Iron

*By R. C. BATES**

Arc welds in nodular iron contain carbide layers at the fusion lines which lower the ductility and impact properties. Postweld annealing eliminates carbides and improves ductility. Oxy-acetylene welds made with nodular iron and Ni Rod 55 are free from carbides, but have high transition temperatures. (K1, K2; CI-r)

SINCE ITS INTRODUCTION in the years after World War II, nodular iron has become an important material for castings. Its properties and economies are attractive to designers. Unfortunately, many applications have been bypassed because available welding techniques did not provide joints with properties comparable to the parent material.

Actually nodular iron can be welded as easily as can gray cast iron since their microstructures differ primarily in the shape of the graphite particles. However, the welds obtained in both materials are generally fragile because brittle carbides form in the weld zone. Post heat treating was reported to be of little value because the extremely small graphite nodules which form upon decomposition of the carbide were believed to reduce ductility.

In most of the early investigations, made shortly after nodular iron appeared, ductility of only 8 to 10% elongation in annealed material was common. Today, nodular iron produced by our Trafford foundry possesses elongations of 18 to 22%. Because of these better properties, we took another look at the welding characteristics.

Welding With An Arc

Arc welding procedures for Ni Rod 55 electrode were first evaluated. (Later, the program included an appraisal of oxy-acetylene welding procedures with nodular iron for a filler metal.) Nodular iron plates supplied by Trafford were 14 in. long, 5½ in. wide and 1 in. thick, and had a composition of 3.68% C, 2.23% Si, 0.041% P, 0.06% Mg, 0.70% Ni and 0.44% Mn. Annealing (for 4 hr. at 1650° F.) and slow cooling produced a ferritic matrix. The mechanical properties (which are typical for Trafford material) were: 62,600 psi. ultimate tensile strength, 47,900 psi. yield strength, and 20% elongation in 2 in.

Welding was done at the Metals Joining Laboratory. To begin with, plates were placed in holding fixtures so that the 30° bevels formed a single Vee with a 60° included angle and a ⅛-in. root opening. With a thin slice of nodular iron as a back-up bar below the root opening, the entire fixture and plates were preheated at 600° F.

Approximately 20 passes with Ni Rod 55 welding electrodes completed each weld. Each pass was preceded by peening and wire brushing. Preheat was maintained at 600° F. with an oxy-acetylene torch, temperature being measured frequently with a contact pyrometer. We found the wetting characteristics of the Ni Rod to be good. Metallographic examination, slow bend tests, tensile testing and impact testing of the plates followed.

Microscopic Examination

Figure 1 shows nodular cast iron as annealed: the structure consists of spheroidal graphite embedded in a matrix of ferrite. A typical weld in the as-welded condition is shown in Fig. 2. Adjacent to the weld metal (austenitic nickel-iron) lies a continuous layer of hard and brittle carbide.

An extremely rapid cooling rate is responsible for this phase. Some of the base metal melts

*Metallurgical Engineering, Materials Engineering Dept., Westinghouse Electric Corp., East Pittsburgh, Pa.

Source: Metal Progress, November 1959

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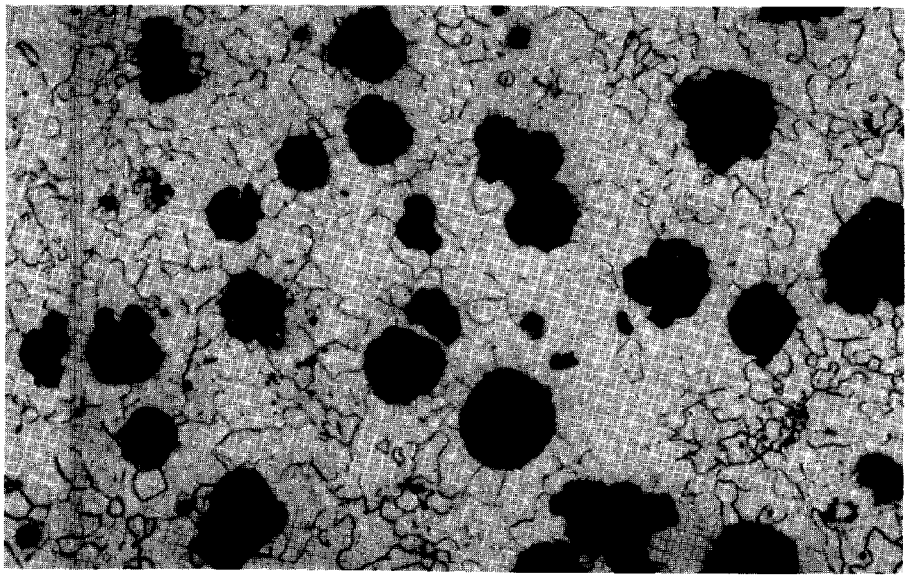


Fig. 1 – Annealed Nodular Cast Iron Structure Consisting of Graphite Spheroids Embedded in a Ferrite Matrix. Etchant, 2% nital; 250 ×

during welding; it freezes and cools rapidly because of the large mass of relatively cool base metal adjacent to the very thin molten layer. Solidification is, therefore, not unlike that which occurs in a chill casting, and the resulting structure is similar to white iron.

Examination of the phase next to the carbide phase at magnifications up to 1500 diameters failed to resolve the structure. Since the hardness is equivalent to Rockwell C-22, the phase probably is spheroidized pearlite. Its properties would be similar to those of lamellar pearlite of equivalent hardness. At greater distances from the weld zone, greater and greater amounts of ferrite are found. Eventually, the structure is completely unaffected, and consists of ferrite and spheroidal graphite.

Annealing at 1650° F. followed with slow cooling decomposes the carbide network and converts the structure of the heat-affected base metal to a mixture of ferrite and nodular graphite. At the fusion lines, however, many tiny particles of nodular graphite are present.

An additional point concerning Fig. 2: The dark phase present in the weld metal is believed to be graphite. Much of it has a flake structure, and probably decreases the ductility of the weld metal. Graphite probably results from the carbonaceous flux which coats the welding electrode.

The tensile and yield strengths of arc weldments of annealed nodular iron are comparable to those of the base metal. The average values found were 60,500 and 44,000 psi., respectively. Elongation in 2 in. was 10%. As-welded tensile specimens exhibit slightly higher strength, but the elongation is only 1 to 2%.

What Bend Tests Proved

Bend tests were performed on unwelded control samples, as-welded samples, and samples annealed after welding. Though most of the samples were tested with the roots of the welds in tension, some as-welded specimens were tested with the weld faces in tension. To aid measurement, lines were inscribed at ⅛-in. intervals on all specimens before testing. Spacing beween these lines was measured after testing.

The unwelded control samples bent an average of 30°. Although most bending occurred near the center line, there was some bending within ¾ in. of each side of the center line. Bending in the as-welded sample amounted to only a total of 2°, and was entirely restricted to the weld metal in an area of less than ⅛ in. on each side of the center line.

No elongation was recorded across the fusion lines. Despite the total lack of strain across these lines (in the as-welded samples), cracks always occurred there. This shows that the continuous carbide network at the weld metal-parent metal interface is brittle and will not support a bending load. Despite this lack of ductility, however, the

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strength of the as-welded joints is equivalent to that of the base metal. Thus, such welds should be considered for applications where only strength, rather than ductility or impact resistance, is needed.

An average bend of 14° occurred during the root bend tests of the arc-welded samples which had been annealed after welding. Most of the elongation was recorded across one of the fusion lines despite the numerous small graphite nodules evident in the microstructure. These tiny particles are apparently not nearly as harmful as previously believed. Also, the ductility exhibited in these tests and the tensile tests is good enough for most applications.

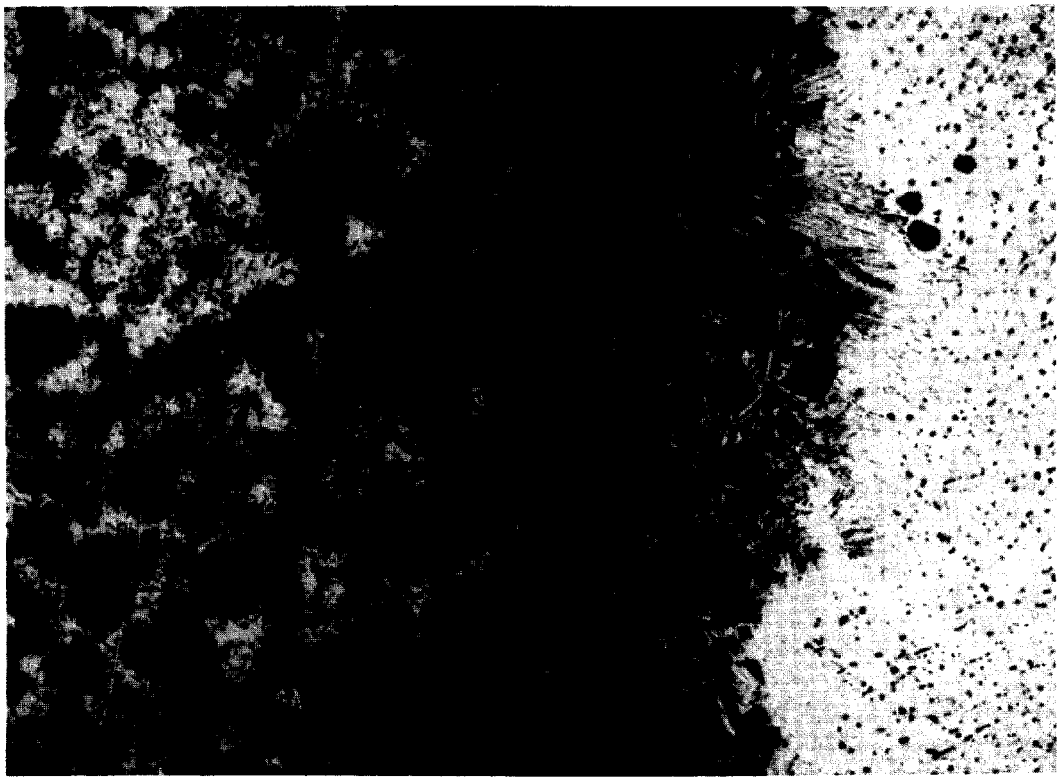


Fig. 2 – Weld Zone in "As-Welded" Nodular Iron. Note white iron (Widmannstätten structure) formed at weld interface. This structure is brittle, and leads to brittleness in the weld. Etchant, 2% nital; 100 ×

Transition Temperature and Nodular Iron

Nodular iron, like many other ferrous materials, exhibits a brittle fracture below what is referred to as the ductile-brittle transition temperature. Actually, the term "transition temperature" is rather indefinite because the transition from ductile shear fractures to brittle cleavage fractures is one which occurs over a relatively wide range of temperature. To evaluate the impact characteristics of the welded and annealed samples, drop-weight specimens were prepared. In this test, a specimen 14 in. long is supported on rests, and a weight of standard size is dropped from a specified height. A notch is provided, usually, by placing a brittle weld bead across the center of the specimen which introduces a running crack after about a 3° bend. A stop prevents bending beyond 5°. Thus, if the specimen breaks during such a test, the mechanism of fracture is primarily cleavage. This is a very brittle type of failure, and complete failure of drop-weight specimens occurs at temperatures below the "nil-ductility transition temperature". The nil-ductility transition temperatures of both unwelded samples and samples annealed after welding were –120 and –95° F., respectively. Thus, arc welding, if followed by an annealing treatment, causes no significant change in the transition temperature. Figure 3 shows the broken half of a drop-weight specimen tested at

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–96° F. Above –96° F., the welded nodular iron samples which were annealed did not break even under the severe impact loading of the test. Since the fracture did not follow the fusion lines in low temperature tests and no fracture occurred above –96° F., the harmful carbide layer in as-welded nodular iron had apparently been eliminated by annealing. The decomposition products left from this carbide layer are not harmful to the mechanical properties, and the resulting joint is adequate for most applications – even those which require impact loading.

No Carbides in Oxy-acetylene Welds

While adequate properties could be obtained in annealed arc weldments, the test results on as-welded joints were not encouraging. Only where no ductility was needed could as-welded joints be used. However, a way out soon appeared. In the *Welding Journal*, September 1957, p. 410, Day, Snyder and Inskeep of the Linde Co. reported that they had successfully prepared carbide-free welds possessing high as-welded ductility. No postweld heat treatment was needed. They did this with an oxy-acetylene welding technique and a specially prepared nodular iron welding rod.

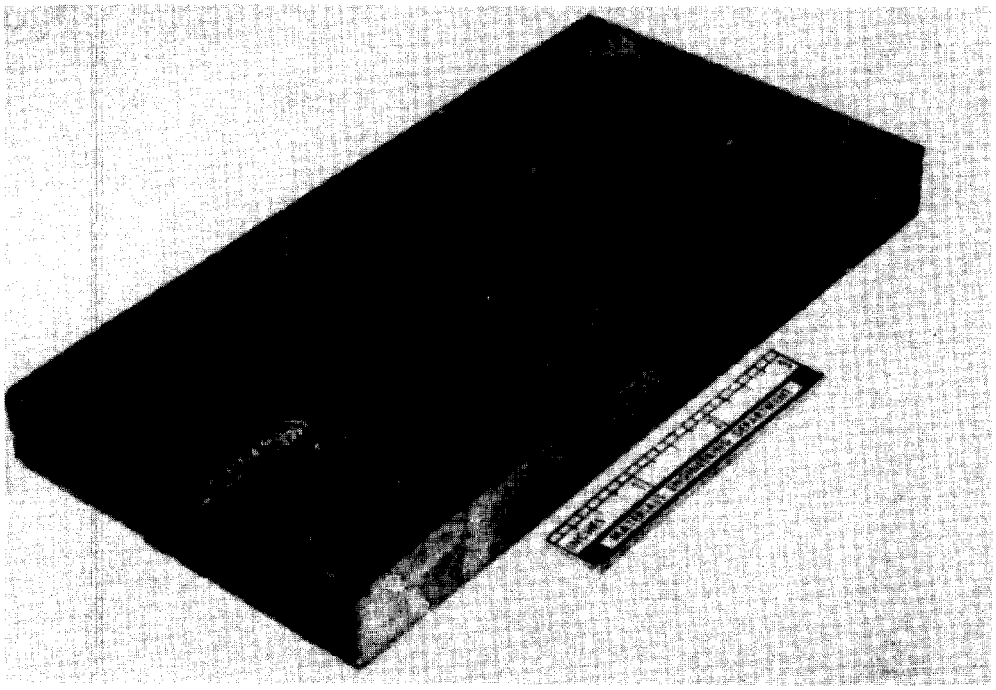
Thus our investigation, originally intended to evaluate only metal-arc welding, was expanded to include the Linde technique. Welding rod and recommendations for welding procedures were obtained from Linde, and plates similar to those previously described were welded.

The samples were placed on firebrick so that the beveled ends formed a single Vee of 60° with a ⅛-in. root opening. (Unlike the setup used for arc welding, no restraining fixtures were used.) First, plates were preheated to 800° F. with the torch. A neutral flame (25 to 30 lb. oxygen and about 10 lb. acetylene) did the welding. Filler metal was Linde Oxweld 69 nodular iron rod, and the weld was fluxed with Oxweld Ferro. Once a puddle was established, successive passes were made continuously until the weld was completed. About six passes were necessary for each weld. After welding, the sample plates were covered with submerged arc flux and cooled slowly to room temperature.

One weld was cracked, but all others were sound except for porosity. Poor wetting in the root area was also encountered, but this was later eliminated with a slightly wider root opening.

Figure 4 shows the interface of the weld metal and parent metal near the root of one of the

Fig. 3 – Welded and Annealed Specimen Which Has Been Tested With a Dropped Weight at –96° F. Brittle failure did not occur in the fusion zone; this indicates that the brittle carbide zone has been removed by annealing



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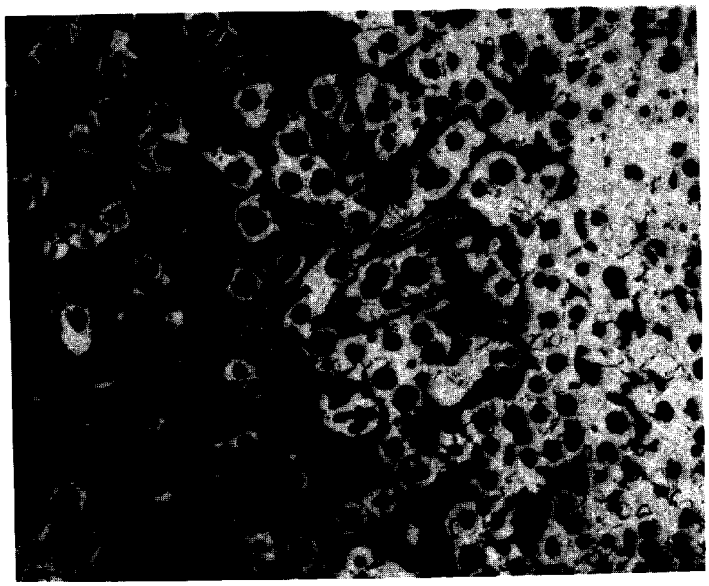


Fig. 4 – Junction Near Root of Weld in Nodular Iron. This weld was made by oxy-acetylene technique with a Linde rod. Note that there is no brittle carbide zone. However, welds produced by this method have high transition temperatures. Etchant, 2% nital. 100 ×

welded joints. The matrix of the weld metal consists of ferrite with small patches of pearlite. Graphite nodules, while smaller than those of the base metal, are spherical. While the interface contains some carbide, the amount is small compared to that contained in the arc-welded samples. In fact, no carbide existed except in the root area, which is illustrated. Other samples examined contained no carbide whatever.

The question arose as to whether it was the special composition of the Linde rod (it contained 3.2% Si) that was responsible for the lack of carbide in the weld zone, or whether the oxy-acetylene welding technique causes slow cooling rates which allow solidification to proceed without carbide formation. To check this, a sample of nodular iron was oxy-acetylene welded with Ni Rod 55 from which the coating had been removed. Welding was difficult because the melting point of the Ni Rod was considerably higher than that of the base metal. An extremely porous weld resulted; however, no carbide was formed.

In addition, we cast some nodular iron at the Trafford foundry. The composition was the same as that of the base plates except that the magnesium content was increased to 0.12% to insure that graphite on the welds would be completely nodular. Welds made with this rod were also free of carbide but the microstructure of the weld metal was primarily pearlitic rather than ferritic.

Since carbide-free welds were obtained with both nodular iron rod and Ni Rod, it is evident that oxy-acetylene welding is a technique which will eliminate the discouraging carbide problem.

Like arc welding, oxy-acetylene welding produces joints in nodular iron with strength comparable to that of the base metal. The ductility of oxy-acetylene welds is somewhat higher than that of unannealed arc weldments (an average bend of 5° occurred during root bend tests). More significantly, the bending was not confined to the weld metal. Instead, bending also occurred across the fusion lines and heat-affected base metal. Thus, the lack of fusion-line carbide allows uniform ductility across the weld zone.

Though this was encouraging, another problem developed. Charpy specimens, machined from the weld joints of several samples, revealed that the nil-ductility transition temperature of welds made with Linde rod is only slightly below room temperature, and the transition temperature of welds produced with Trafford rod is above room temperature. Thus small defects may trigger premature failures.

At the present time, while the carbide problem has been eliminated, the problem of high transition temperature remains. This problem must be eliminated before oxy-acetylene welding may be recommended for general use in joining nodular iron. Such weldments may be used, however, if the design stress is low enough to insure that the yield strength is not exceeded (even in the vicinity of stress-raisers), or if adequate inspection can insure freedom from defects which may act as severe stress-raisers. Oxy-acetylene welding may be particularly useful in the future for repairs in noncritical areas where postweld annealing is not permitted. ●

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New Ductile Iron Welding Process... Saves Castings and $

by D. T. Roberts,
Manager Foundry Operations
Oil City Iron Works
Corsicana, Texas

Now you can repair or assemble ductile iron castings by a new gas welding process developed by a Texas foundry.

ANY METALCASTER making ductile iron recognizes the need for a simple, reliable welding technique that will aid in salvage repairing and joining of castings. We felt the pressure of this need ever since 1960 when we started casting ductile iron at Oil City Iron Works. So we embarked on a research and development program of our own to meet this need.

After four years of development, and examination of hundreds of weld specimens, we can now announce a ductile iron welding capability with:

1) Perfect color match.
2) Machinability at normal speeds without heat treatment.
3) Similar chemical composition.
4) Similar or identical microstructure.
5) Similar or identical physical properties.
6) Response to heat treatment in predictable manner.
7) Minimum transition zone.
8) Minimum heat affected zone.
9) No excessive sensitivity to technique.
10) Easy field welding.
11) Minimum stresses developed.

Yes, improved production methods, available alloys and greatly advanced techniques have reduced plant rejects to a practical minimum. But where permissible, this welding process can be used to repair many of the few remaining rejects, increasing casting yield further. Also, ductile castings can now be joined by a practical gas welding process, opening up a new field of applications. As ductile iron markets continue to expand the process becomes even more valuable and profitable.

Here's How We Did It

You need this equipment (all displayed in Fig. 1):

1) Suitable operator protective equipment
2) Acetylene gas and oxygen with appropriate pressure regulators
3) Grinding or scarfing devices for preparation of area to be welded and finishing welded area
4) Burner (natural gas or butane) for preheat of part
5) Powder spray type oxyacetylene torch fitted with appropriate tip.

The area to be welded *must* be clean and free from scale, rust or other impurities. This may be accomplished by any convenient means such as chipping and/or grinding. Normal good weld cleaning procedures are adequate. Conventional cleaning room equipment should prove satisfactory for this operation.

In joining two members, each member should be notched at the point of fusion welding in conventional fashion.

Preheating of the area to be welded serves two desirable purposes:

1) Retards post welding cooling rate to avoid presence of cementite or martensite in the transition zone
2) Avoids or minimizes excessive casting stress and/or strain.

Local preheating of area (Fig. 2) to be welded with methane or LP gas torch to temperature range of 500F to 700F is sufficient to accomplish these goals. Complicated castings may require more elaborate preheating.

For fusion welding a slight excess of acetylene is used to insure a slightly reducing flame. The flame is brought in contact with the area to be welded. When melting begins, the rod is introduced into the flame and welding by puddling progresses in the usual manner. Puddling may be accomplished by either backhand or forehand tech-

Source: Modern Casting, February 1969

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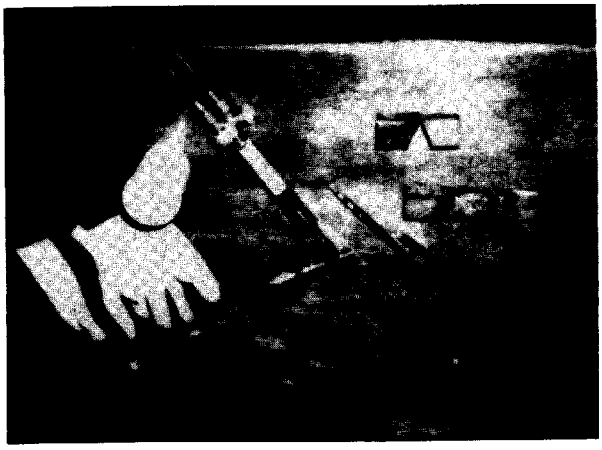


Fig. 1. Display of required equipment.

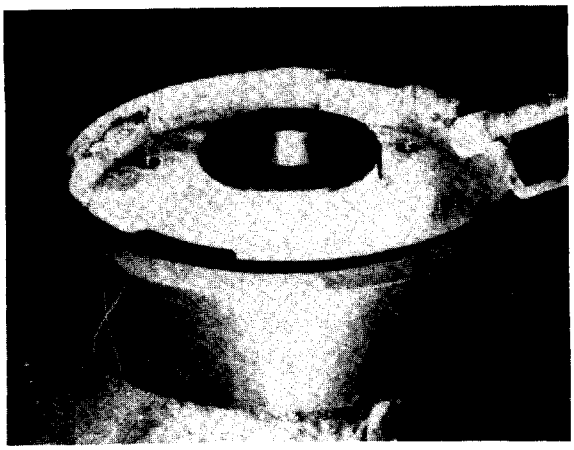


Fig. 2. Preheating of part prior to salvage via welding.

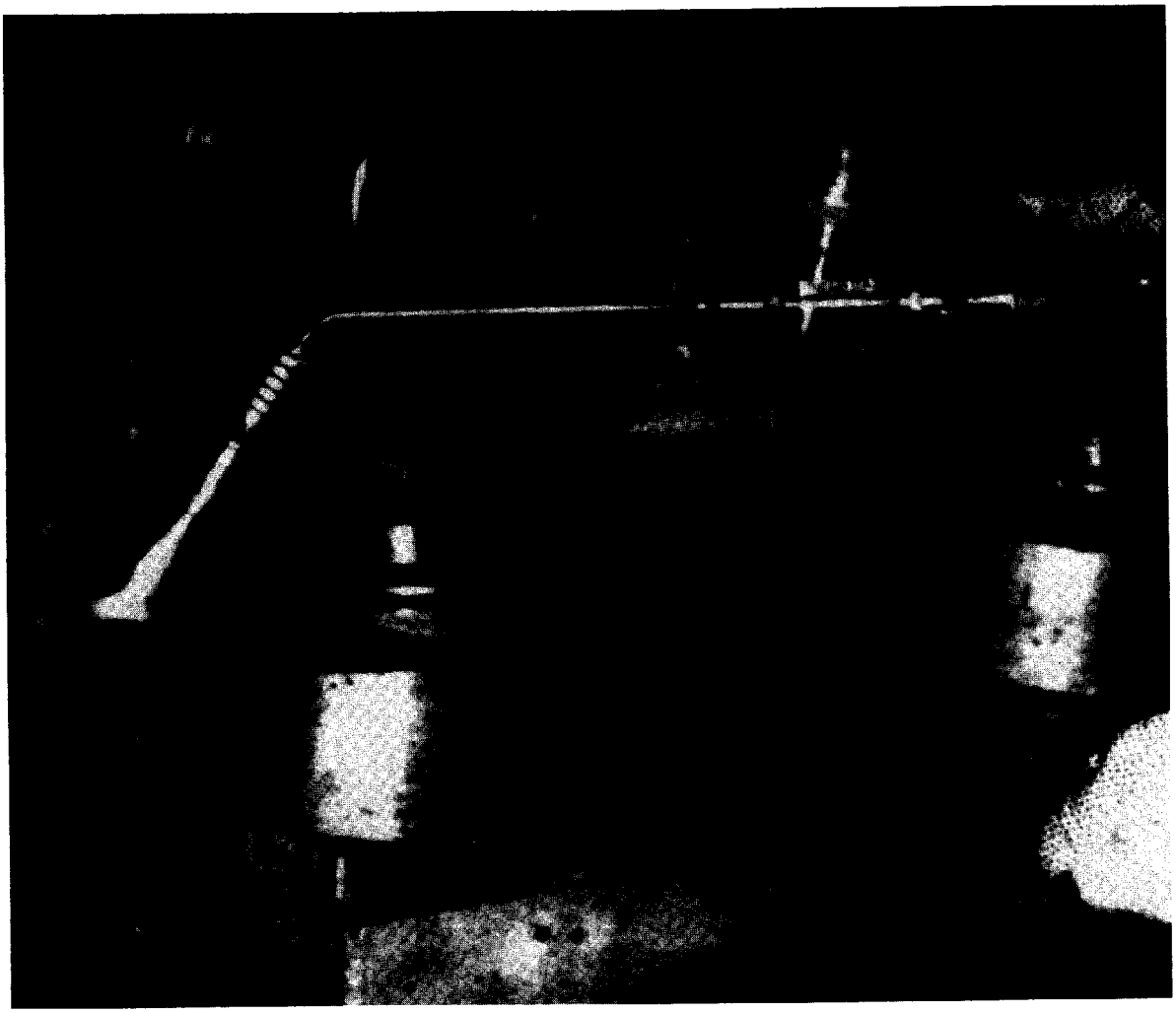


Fig. 3. Salvage welding of medium size ductile iron casting.

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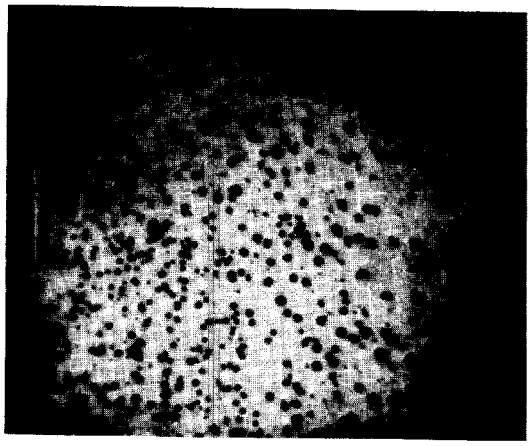


Fig. 4. Transition zone. Note difference in nodule count of parent metal (right) and weld deposit. 100×-Unetched.

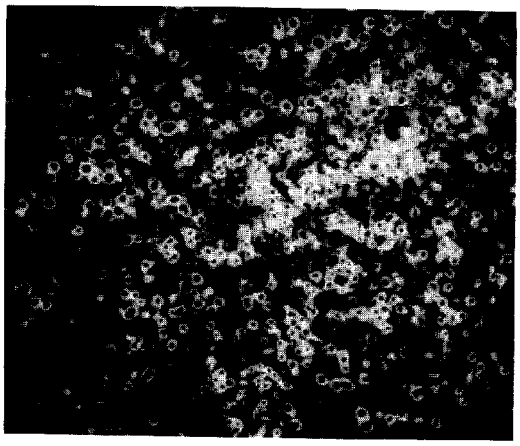


Fig. 5. As welded-100×-2% nital etch. Matrix of pearlite and ferrite.

niques with the latter preferred in that the rod leads the progression of the weld. Flux of special composition is added to the puddle through the flame in normal powder spray fashion.

The flux is so adjusted as to promote a fluid slag cover necessary to sound gas welds as well as to insure complete nodularity of the weld deposit. A patent for the process is pending.

The cooling rate of the completed weld deposit may be further retarded by continuing to play the torch on the weld deposit after solidification. Post heat treatment is not required to accomplish a free machining welded area with excellent "color match". The welded area will respond to conventional heat treatments such as ferritizing, normalizing and quench and temper in a manner like the parent metal (assuming normal chemical composition). Carbide stabilizing elements will affect response to heat treatment.

Results of ASTM 1-in. Y-Blocks from Single 1200-lb Ladle.

Bar No.	Condition	Tensile psi	Yield psi	El %	Bhn
As Cast					
214	Unwelded	87,600	68,400	8.0	217
214-A	Welded	84,400	67,200	5.0	223
214-B	Welded	86,200	67,800	6.0	223
Annealed					
215	Unwelded	65,200	47,800	19.0	156
215-A	Welded	66,400	46,000	16.0	149
215-B	Welded	62,600	48,200	14.5	143
Normalized					
216	Unwelded	93,400	76,400	4.0	235
216-A	Welded	97,200	77,000	4.0	241
216-B	Welded	100,200	78,200	3.5	255
Quenched and Tempered(1000F)					
217	Unwelded	121,600	90,800	2.5	302
217-A	Welded	117,200	87,200	3.5	269
217-B	Welded	120,200	91,400	2.0	293

Only six tests of physical properties have been performed on ductile iron welded to 1020 steel with encouraging but inconclusive initial results as follows:

					Red. of Area-%
138*	As welded	53,850	38,850	23.0	67.3

*Failure occurred in steel parent metal.

Due to a more rapid cooling rate of the weld deposit than is probable with the parent metal, the nodule count of the weld deposit will be higher than that of the parent metal. (See Fig. 4)

A properly executed weld will be essentially free from cementite (iron carbide) in the weld deposit or transition zone. It will possess a pearlitic or sorbitic matrix with a small bull's eye of ferrite around the graphite nodule—Fig. 5.

Conclusions

1) The process described is practical under normal operating conditions and not a laboratory technique.
2) It is possible to salvage defective ductile iron castings, before or after machining by oxy-acetylene welding with resulting weld deposits similar to parent metal in chemical analysis, microstructure and physical properties.
3) Ductile iron parts which have been scrapped through machining error can be salvaged.
4) Complex parts may be produced by joining of several simple cast parts at lower cost than casting as integral part.
5) Ductile iron castings may be repaired or joined in the field since no heat treatment is required.

For more information about this welding process circle No. 570, on last page.

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High Strength Welding of Ductile Iron Castings

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ABSTRACT

After a short review of problems arising during the welding of ductile iron castings, the present paper deals with development of a weld material suitable for obtaining high strength welds. Such a material basically consists of cast iron enriched with additives that will act during solidification of the joint to cause graphite to segregate in form of spheroids, without any lamellar flakes. The joint so produced is then free from the microstructural lack of homogeneity that occurs with usual welding processes. Accordingly, the mechanical properties of the welded joint are similar to those of the base material.

Some steps in development of such a weld material are described. The target of getting the separation of carbon (C) only in the spheroidal form was achieved by replacing the usual spheroidizing element, magnesium (Mg), with cerium (Ce) and rare earths, which have far higher boiling points than (Mg), and therefore can effectively play their spheroidizing role during the melting of the welding bed.

The influence of nickel (Ni) and silicon (Si) in the weld material was also evaluated. Moreover, welding rods with a reduced rare earth content and with copper (Cu) instead of Ni were investigated and the results reported. Welding between ductile iron castings and steel pieces, like forged or laminated parts, was also investigated. The high mechanical properties obtained in the welded areas confirm the feasibility of this type of weld and allow a wide range of industrial applications to be considered.

Introduction

Cast iron is generally considered a material not easily weldable and difficulties are especially great with nodular cast iron because of its high C content. During solidification of the melted welding zone, a part of the C tends to separate in form of lamellar graphite while another part combines with the base material giving rise to stable carbides. Therefore mechanical weld properties are considerably lower than those pertaining to the base material.

These problems have been overcome. This paper deals with the development of a weld material which can be used in nodular cast iron welding and is suitable for attaining high strength welds. Actually, by using normal techniques and care as needed for good welding, it is possible to produce uniform weld free of defects, in which the same structural characteristics and mechanical properties of the base material are present.

After a wide practical evaluation in the field, this material has been currently employed for years at an important foundry, for welding needs on heavy and medium weight ductile iron castings, mainly for tractors and trucks.

The Problem of Ductile Iron Welding

Ductile iron, as well as gray iron, is usually welded using an oxy-acetylene torch or an electric arc method. Weld materials in the forms of rods or electrodes may differ in form, section and chemical composition and the choice of the right types depends on weld needs and purposes.

Cast iron-base materials are employed when welding cast parts for removing defects such as blowholes, cracks and porosity by simply filling them and does not aim at obtaining high mechanical properties. Therefore, this method of welding is only used for coarse recovery operations in stress free zones. Cu-base rods employed in brazing are also used for this purpose. Deformation and cracks are therefore less likely to occur, as the brazing temperature never exceeds 1830F (about 1000C).

Alloy materials, mostly Ni-based, are used to meet more qualified welding requirements. In this case welding is characterized by improved mechanical properties: ultimate tensile stress in a range of 70,000 to 85,000 psi (about 50 to 60 kg/mm^2) and yield strength of 57,000 psi (about 40 kg/mm^2) have been attained by means of suitable heat treatment. However, it must be remembered that such remarkable strength represents the upper limit of the weld material, but the welded joint cannot be considered homogeneous either as a microstructure or in mechanical properties.

Microstructural Considerations

During welding, various sections of a casting are heated to temperatures slightly higher than temperatures at which the part was originally cast. The material is consequently subjected to structural changes like those which take place during heat treatment.

The diagram in Fig. 1 graphically shows structural changes which occur transversally to the welding bed in hypereutectic ductile iron during welding. The extent of these changes is related to the temperature gradient present in the castings during welding.

The final structure of the welded casting may be characterized by remarkable differences, which depend on cooling modalities. The diagram in Fig. 1 clearly shows that material in the heat affected zone changes from a pearlitic-ferritic structure with free graphite before welding, to an almost austenitic structure. Later on, during cooling which is usually rapid, austenite turns into a pearlitic structure finer than the previous one with ferrite. The ferrite is now almost absent or it turns into a structure ranging from martensite to bainite in limited zones where the critical cooling rate was overcome.

In the melted zone, structures clearly depend on the weld material used. When cast iron base weld materials are used, cementitic structures easily originate; the rapid cooling prevents the ledeburite and cementite decomposition (a higher percentage of Si content and a lower cooling rate help to reduce this inconvenience). Part of the C separates in the form of lamellar graphite, even if the weld material composition is of a common ductile iron.

When Ni-base weld materials are employed, comparatively stable martensite structures originate in the transition zone owing to the chemical composition, thus reducing the critical

Table 1. Mechanical Properties of Base Material

	As cast	Normalized
UTS psi (Kg mm^{-2})	93,000 (65)	128,000 (90)
0.2Y.S. psi (Kg mm^{-2})	68,000 (48)	74,000 (52)
E %	5	5
Bhn	240 + 280	250 + 300

Source: AFS Transactions, 78-86, 1978 (American Foundrymen's Society)

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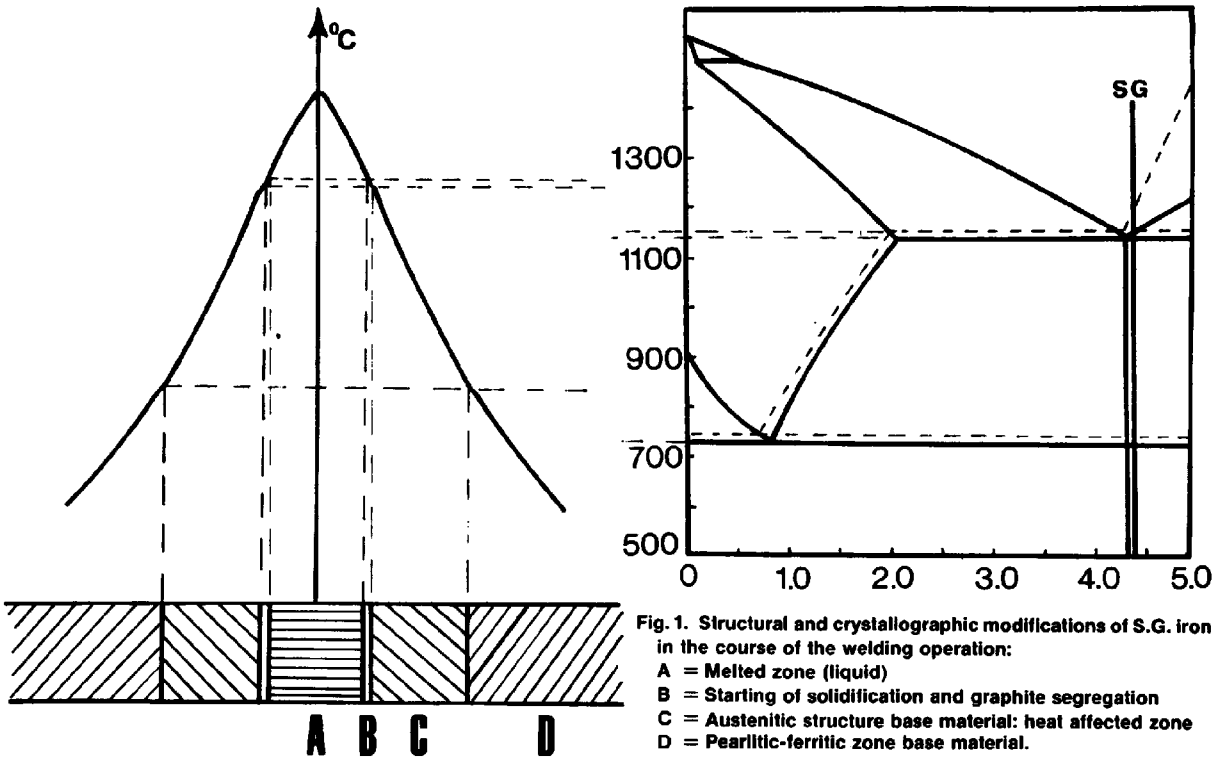


Fig. 1. Structural and crystallographic modifications of S.G. iron in the course of the welding operation:
A = Melted zone (liquid)
B = Starting of solidification and graphite segregation
C = Austenitic structure base material: heat affected zone
D = Pearlitic-ferritic zone base material.

cooling rate. Nickel from the weld material, C and Fe from the melted base material are also present in this zone in different percentages (otherwise, without any diffusion a brazing would occur instead of welding). It must also be observed that an extremely hard and brittle structure may originate when Ni content is higher than 50%.

As regards the graphite morphology, considerations similar to those mentioned above can also be applied to these Ni-based materials, i.e., the C amount separating in the form of graphite is smaller, because it only originates from the base material, but it turns into a lamellar shape, thus reducing mechanical properties.

Improvement of Ductile Iron Welded Joints

Micrographic examination of ductile iron joints welded by Ni-base rods or electrodes shows a discontinuity and heterogeneity of the micrographic structure. This fact must be overcome if a remarkable improvement is to be made. To this purpose a weld material capable of causing graphite separation in the form of nodules during solidification has been studied and developed.

Such a material, covered by patent,[1] basically consists of cast iron characterized by a composition similar to that of normal nodular cast iron, but enriched with additives, which act during the solidification of the joint to cause graphite separation in the form of nodules.

Experimental Tests

The properties of materials used in the experimental stage and the kinds of tests carried out on welded castings are reviewed before dealing with weld material development.

Base Materials

A common pearlitic nodular cast iron (corresponding to the ASTM A 536-70 grade 80 - 55 - 6) was used as base material for the preparation of different test welds: the chemical composition of this cast iron was as follows.

C	Si	Mn	S	P	Mg
3.60%	2.50%	0.40%	0.010%	0.05%	0.03%

The mechanical properties of this iron, in the as cast and normalized states are shown in Table 1. The relevant fatigue limits (at 5 x 10^6 cycles) to alternate plane bending are shown in Table 2.

Table 2. Fatigue Limits of Base Material

		As cast	Normalized
R = -1	psi (Kg mm^{-2})	43,000 (± 30)	43,000 (± 30)
R = 0	psi (Kg mm^{-2})	36,000 (± 25)	36,000 (25 ± 25)

Heat Treatment

Some heat treatments have been carried out after welding:

a) Carbide removal
- heating up to 1650F (900C)
- holding at 1650F (900C) for complete austenitizing
- rapid cooling down and holding at about 1300F (700C) giving a ferritic-pearlitic structure.

b) Carbide removal normalizing
- heating up to 1650F (900C)
- holding at 1650F (900C) for complete austenitizing
- controlled cooling down, giving a uniform homogeneous pearlite.

The carbide removal cycle is always required on test pieces,

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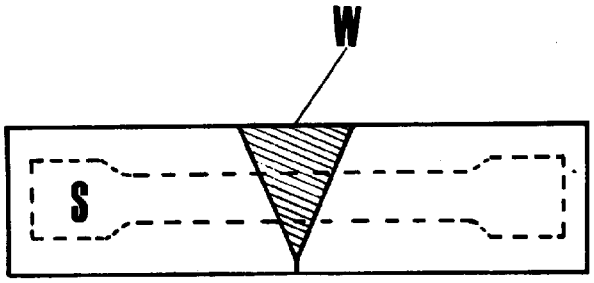


Fig. 2. Specimen for the determination of tensile strength obtained from a welded bar:
S =Specimen
W =Welding.

because they cool rather quickly due to their low thermal capacity.

Tests Made on Welds

The following tests were carried out on welded specimens:

a) Static mechanical properties:
Standard tensile specimens were obtained from welded bars, Fig. 2; particular attention was given to placing weld areas in the middle of the sample's length
b) Brinell Hardness:
Measures were taken transversally to the weld before and after heat treatment
c) Fatigue limit:
Fatigue limits to alternate plane bending were measured on specimens obtained from properly dimensioned keel blocks, Fig. 3
d) Metallographic inspection of the added material:
Micrographic inspection was carried out on suitable specimens and on tensile test pieces starting from the heat unaffected base material up to added material melted.

Weld Materials and Results

Rods characterized by the same composition of common ductile iron and with Mg contents ranging from 0.04 to 0.10% were first tested in order to carry out the spheroidizing treatment directly in the melted zone.

In all cases results were rather poor, both from the micrographic point of view, Fig. 4, which showed the presence of lamellar graphite and from the viewpoint of mechanical properties as shown in Table 3.

Magnesium concentration on the weld was assessed and it proved to be very low (i.e. about 0.005% against 0.030% of the base cast iron). This Mg loss has to be ascribed partly to

Table 3. Mechanical Properties of S.G. Joints Welded with Welding Rods Having Different Contents of Mg

Mg content %	UTS psi (Kg mm-²)		E %	
	as cast	normalized	as cast	normalized
0.04	42,800 (30)	41,400 (29)	1	0.4
0.10	51,400 (36)	48,500 (34)	1	0.3
Base material	85,700 (60)	128,600 (90)	5	5

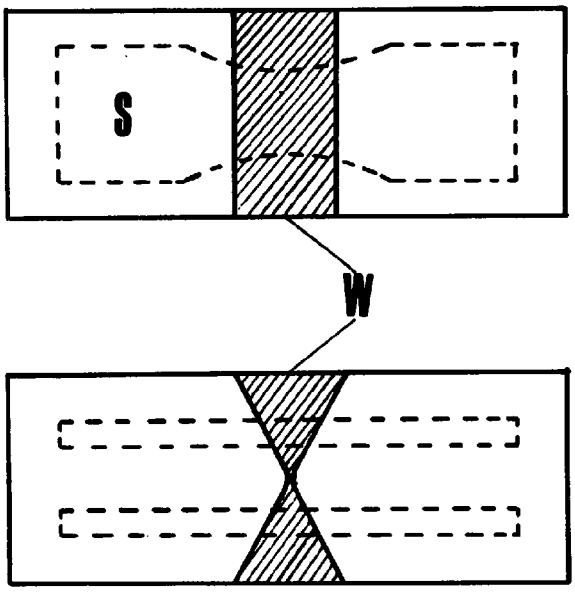


Fig. 3. Specimen for the determination of fatigue limit obtained from properly dimensioned welded keel block:
S =Specimen
W =Welding.

oxidation and partly to evaporation because of the high temperature: therefore, spheroidizing does not take place because of an insufficient concentration of the spheroidizing element.

The loss as a result of oxidation could be eliminated by carrying out the welding operation under inert atmosphere, but evaporation loss would still exist; in fact the boiling point of Mg is 2042F (1117C), lower than the temperature present in the zone melted during welding. This trouble was overcome by substituting Mg with a less volatile spheroidizing agent such as Ce and other rare earths.

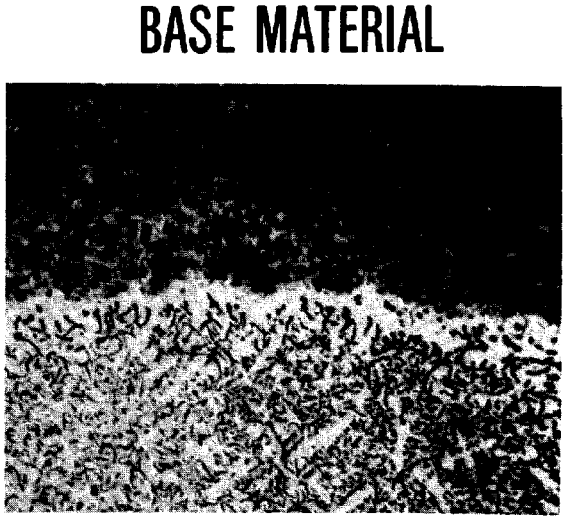


Fig. 4. Microstructure of S.G. iron welded with 0.1% Mg rod. (100X)

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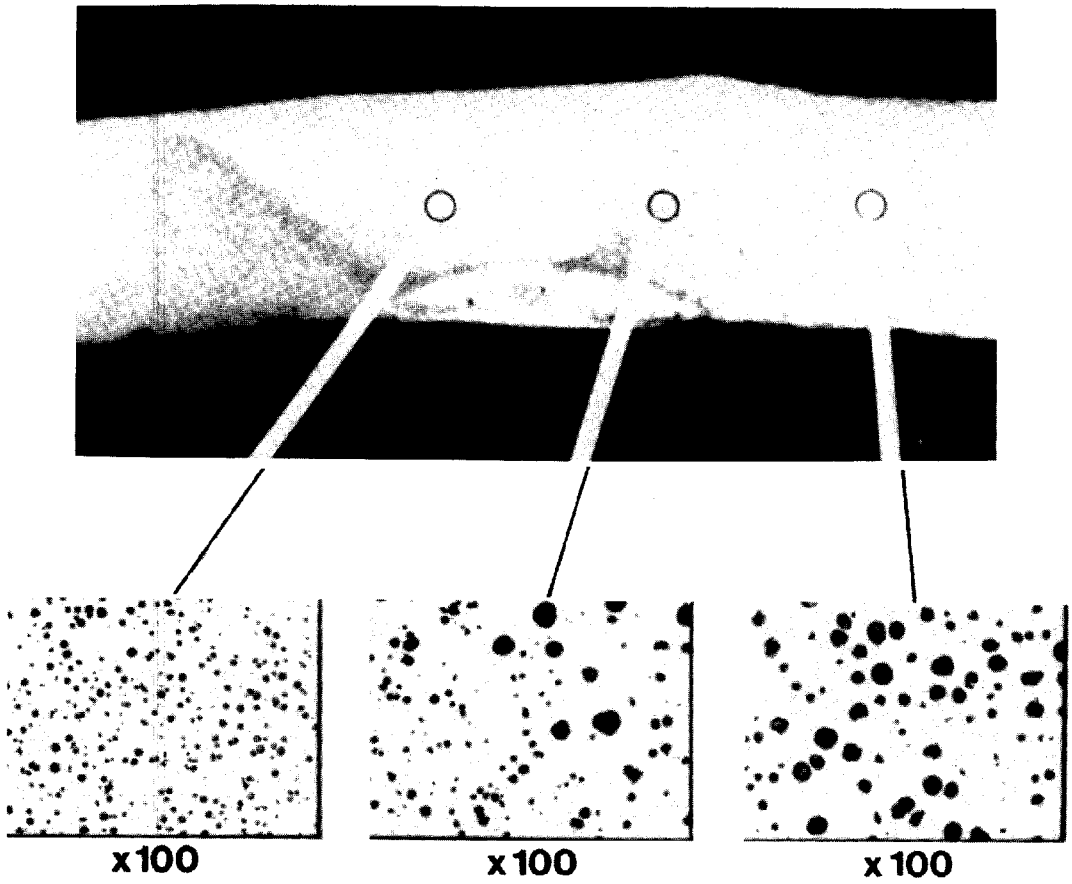


Fig. 5. Macrograph of a welded joint and micrographs of typical areas.

From the literature it is known that the boiling point of these elements is much higher than that of Mg, i.e., Ce about 5306F (2930C), La about 6098F (3370C), Pr about 5450F (3017C) while they are all characterized by a melting point below 1830F (1000C). In spite of some losses as a result of oxidation, a good spheroidization could be achieved in the welding bed by using a weld material containing a spheroidizing element of this type. Different kinds of rods for oxy-acetylene welding were then designed, manufactured and tested according to these concepts.

The effects of Si content were taken into account and its percentage was metered in such a way as to get a Si content in the bed almost equal to that which can be found in the base cast iron. The Si percentage in the rod material has been increased in order to compensate for Si oxidation losses, which are about 30% in a standard welding cycle. Silicon increases metal fluidity and therefore simplifies deposition of the metal. Besides, it promotes cementite transformation to pearlite and graphite during cooling. Too high a content would, however, decrease the weld tensile strength.

The effect of Ni has also been evaluated. Its amount was maintained at a level high enough to increase weld plasticity and thereby to decrease possibility of crack formation, but insufficient to cause formation of brittle structure such as (Fe Ni) C_3.

On the basis of these considerations, rods for oxy-acetylene welding 12 in. long (300 mm) and 1/4 in. dia. (6 mm) were cast

Table 4. Tensile Properties of Base Materials and of Welded Joints

	UTS psi (Kg mm^{-2})	YS psi (Kg mm^{-2})	E %
Base material			
as cast	92,850 (65)	68,500 (48)	5
normalized	128,500 (90)	74,300 (52)	5
Welded joint			
perlitic — ferritic	88,500 (62)	59,950 (42)	3.5
normalized	114,200 (80)	82,750 (58)	2

Table 5. Fatigue Limits, at 5 x 10^6 Cycles, Alternate Bending

	R = -1 psi (Kg mm^{-2})	R = 0 psi (Kg mm^{-2})
Base material		
as cast	± 42,830 (± 30)	35,680 ± 35,680 (25 ± 25)
normalized	± 42,830 (± 30)	35,680 ± 35,680 (25 ± 25)
Welded material	± 37,100 (± 26)	32,830 ± 32,830 (23 ± 23)

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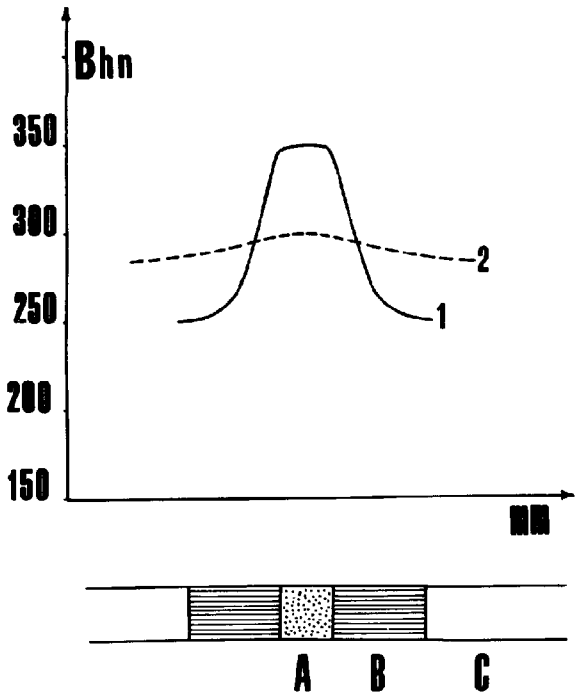


Fig. 6. Hardness values measured transversally to the welded zone before and after heat treatment:
A =melted zone
B =heat affected zone
C =base material
1 =as welded
2 =normalized.

according to the following composition.

C	Si	Ni
3 to 3.2%	3.3 to 3.5%	4.9 to 5.1%

Mn	Rare Earth	S	P
0.3 to 0.4%	0.5 to 0.8%	0.01%	0.04%

Test welds, as previously described, were made using these rods. Micrographs from different zones of a butt welding (base material, transition zone and weld material) are shown in Fig. 5. Figure 6 shows hardness measured transversally to the welded point before and after heat treatment for carbide removal.

Data about mechanical weld properties are shown in Tables 4 and 5. The first series of data was obtained from the fracture of tensile specimens, while the second series concerns fatigue tests to plane alternate bending. The data obtained show that mechanical properties, both static and dynamic, are very close to those of the base metal. Micrographs and data are in agreement with the excellent mechanical and structural quality that can be attained.

This kind of rod enables a good spheroidization to be obtained, which tends to be fine and uniform because of the relatively high solidification rate. Such a structure guaranties mechanical characteristics as good as those of the base material; the only condition being that welding is done properly to produce a joint free of welding flaws.

Moreover, rods with different chemical compositions have been produced and tested; the rare earth content was reduced and Ni was replaced by Cu. It was observed that the

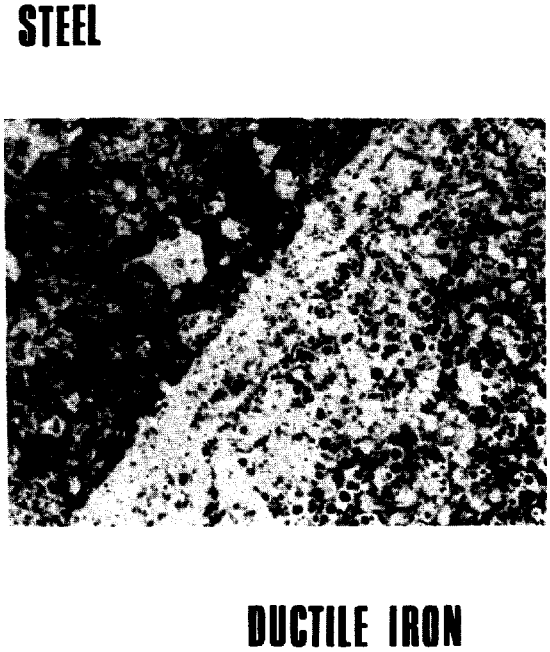


Fig. 7. Microstructure in a welded joint between S.G. iron and steel. (100X)

spheroidizing element can be reduced to 0.4% without affecting the quality of the spheroidization. The partial or total Ni replacement by Cu still enables us to obtain a high tensile strength with a modest elongation reduction.

As regards welding modalities, the same care used with other materials for oxy-acetylene welding is required. The condition of surfaces to be welded is very important. Foreign material such as oil, grease, paint, rust, sand and earth must be removed, as these impurities can cause inclusions and blowholes.

The edge preparation can be done using the same criteria usually adopted for welding steel. We must in particular mention butt joints carried out with bevel angles ranging from 40° to 90°; easy uniform operation and good mechanical characteristics were obtained in all cases. It should be pointed out that the practical performance of the welding operation turns out to be easy and quick thanks to the favorable combination of the surface tension and fluidity found in the weld material.

Complete casting preheating is not generally necessary. Local preheating with the torch is sufficient just before welding. Preheating at about 600C (1112F) is only advisable for castings with very irregular mass distribution, where dangerous stresses might arise. The decrease of the temperature gradient prevents possible crack initiation. In order to eliminate carbides and internal stresses, some form of heat treatment is generally needed as in any other process.

Welding Ductile Iron With Steel

In order to verify the welding feasibility between heterogeneous materials, welds between ductile iron and SAE 1040 and SAE NE 9837 steels were made using rods containing 5% Ni, 0.8% R.E.

The micrographs of Fig. 7 show the structure in the welded zone. Carbon diffusion phenomena towards steel occur. Conse-

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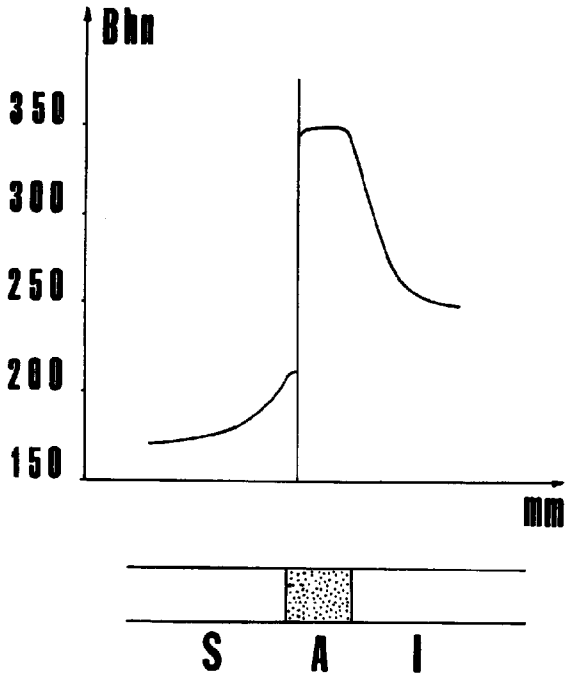


Fig. 8. Hardness values measured transversally to the welded zone between S.G. iron and steel:
A =Melted zone
S =Steel
I =Ductile iron.

quently, a gradual appearance of nodules of free graphite characterizes the interface. Then the structure turns into a proper steel structure through a martensite zone. Hardness figures measured on the transverse welding section, are shown in the diagram of Fig. 8. As expected, the highest value was found on the weld zone.

Table 6 shows the mechanical properties found on tensile specimens obtained from the weld zone. The specimen fracture always takes place in the ductile iron side at an average distance of .1 in. (2 mm) from the interface with the added material, and therefore in the heat affected cast iron zone.

Table 6. Mechanical Properties of Welded Joints Between S.G. Iron and Steel

Base materials	UTS psi (Kg mm^{-2})	YS psi (Kg mm^{-2})	E %
S.G. iron and SAE 1040 steel	82,000 (58)	61,000 (43)	3
S.G. iron and SAE NE 9837	90,000 (63)	77,000 (54)	2.6

The results obtained show that ductile iron can be effectively welded to steel by using the proposed material.

Conclusion

According to the results reported in this paper, it can be stated that ductile iron welds achieve a high structural homogeneousness and high mechanical strength. At this time, only tests on oxy-acetylene welds have been completed as the applicable technology permits use of rods with diameters greater than 5 mm (0.175 in.).

Further work is now being performed in order to obtain rods with a smaller diameter, 1/8 in. (3 mm), with the same type of material. These rods are intended for use as coated electrodes in arc welding.

The fact that high-resistance welding on ductile cast iron using the new weld material can be carried out represents an important improvement, even if, at present, limited to oxy-acetylene welding. This method is most advantageous both as a salvage operation of expensive castings and for production of assembled units composed of several elements made of ductile iron.

A deeper knowledge of this type of welding may lead to a greater flexibility in designing weld-assembled components including ductile iron castings and steel sheets or forgings, so that we may obtain full benefit of each technology and thereby obtain the most economical method of producing the job.

Reference

1. E. Natale, Italian Patent No. 68062A/77 (May 11, 1977).

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Secondary Graphite Formation in Tempered Nodular Cast Iron Weldments

Graphitization of martensite and carbides during postheat treatment produces secondary graphite nodules in the heat-affected area; preheating can control the amount and morphology of these nodules

BY D. R. ASKELAND AND N. BIRER

ABSTRACT. Postheat treatments below the critical temperature are effective in reducing the hardness of nodular cast iron weldments by eliminating carbides and martensite. However, because of the precipitation of secondary graphite nodules, such postheat treatments may not completely restore ductility to the weldment. High preheat temperatures minimize martensite and the amount of secondary graphite that forms during tempering; they also increase the amount and continuity of the carbides, resulting in elongated and chain-like graphite particles.

Full annealing of the weldments prevents the formation of chain-like graphite but does not completely prevent the formation of secondary graphite nodules. Best results might be obtained by combining a suitable preheat temperature to prevent martensite formation with proper selection of base metal microstructure and welding process and parameters to minimize carbide formation. A full anneal gives better microstructure control than a subcritical heat treatment.

Introduction

The microstructure of nodular cast iron weldments is characterized by the formation of hard, brittle iron carbides and high carbon martensite in the heat-affected area. Although these constituents can subsequently be eliminated by a lengthy heat treatment, secondary graphite particles are produced which still impair the ductility and impact properties of the weldment.

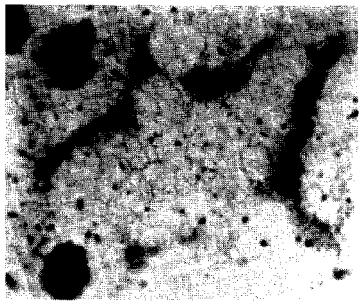


Fig. 1—Round secondary graphite nodules and uncoalesced carbon clusters in the heat-affected region of a nodular iron weldment after tempering at 1250 F (677 C). X500 (reduced 50% on reproduction)

In nodular iron weldments, the secondary graphite particles originate from two sources. First, iron carbide forms in the fused and diluted base metal adjacent to the fusion zone due to fast cooling rates and the absence of effective nuclei for graphite during solidification. The iron carbide is difficult to prevent except by using extremely high preheat temperatures,[1] although the continuity of the carbides can be disrupted by control over the initial microstructure and the welding parameters.[2] When the nodular iron weldment is postheated at very high temperatures, the carbides will graphitize, just as white cast iron is converted to malleable cast iron, producing small graphite nodules from the carbides. These nodules may often be arrayed in the form of long chains of graphite particles.[3]

The second source of secondary graphite is high carbon martensite located in the partial fusion zone and transformation zone of the weldment. Large amounts of carbon can be dissolved in austenite during the welding cycle by solution of carbon from the original primary graphite nodules. On rapid cooling, high carbon martensite will form. Initially, clusters of tiny carbon particles form from the martensite during tempering; the clusters eventually coalesce into distinct secondary graphite nodules.

Figure 1 shows an example of both uncoalesced carbon clusters and secondary graphite nodules in the heat-affected region of a tempered weld. These secondary graphite particles originate at lattice defects, particularly at martensite plate boundaries or intersections[4,5]; consequently, their number is at least partly determined by the structure and coarseness of the martensite. The secondary graphite is only observed after tempering at or above about 1000 F (538 C) and after the martensite has at least partially ferritized.[6-8] The amount and morphology of the secondary graphite depends on the amount of carbon in the martensite, or on the peak temperature and time to which the austenite was exposed.

Higher austenite temperatures or

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longer times at elevated temperatures increase the amount of carbon dissolved in the martensite. For low carbon martensite, no secondary graphite forms; for intermediate carbon levels the secondary graphite is round and randomly distributed, while at high carbon contents elongated nodules or chains of secondary graphite are produced.[8] Elongated ferrite grains are also normally associated with the high carbon contents.

Procedure

Nodular iron plates 0.5 in. (1.27 cm) thick were poured in green sand molds after the iron was melted in a 50 pound (22.8 kg) capacity magnesia-lined high frequency induction furnace. The iron was nodulized at 2710 F (1488 C), inoculated, and poured at 2550 F (1400 C). The composition of the cast plates was 3.6% C, 2.14% Si, 0.39% Ni, 0.038% Mg, and 0.013% S. The high nickel content helped assure that large amounts of martensite would form in the heat-affected area during cooling of the weld, both by promoting pearlite in the base metal of the casting and by increasing the hardenability of the austenite.

Castings were preheated in an electric muffle furnace; immediately after removal from the furnace, a bead-on-plate weld was produced using the shielded metal arc welding process with 0.125 in. (0.3175 cm) CI-Ni electrodes. The heat input varied from 27,000 to 44,000 joules/in. (2540 to 4140 cal/cm). Samples were cut from the welded plate and heat treated in an electric muffle furnace for up to sixteen hours. Metallographic specimens were prepared from the welds and etched in 5% nital.



Fig. 2—Typical microstructure of a nodular iron weldment with no preheat, showing the carbide, partial fusion, and transformation zones. X250 (reduced 50% on reproduction)

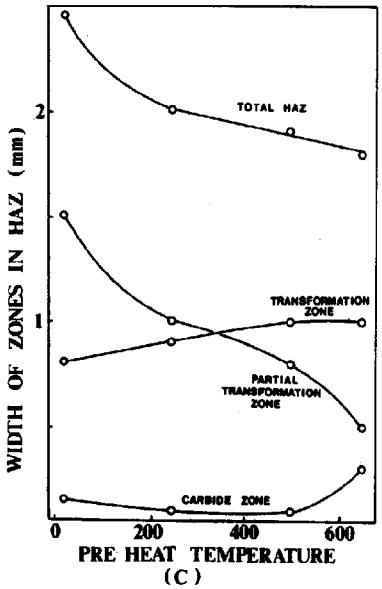


Fig. 3—The effect of preheat temperature on the width of the zones in the heat-affected area of nodular iron weldments

Results and Discussion

Microstructure of Weldments

Figure 2 shows a typical microstructure of an as-welded nodular cast iron plate, with no preheat. Five regions in the weld were identified to simplify subsequent analysis:

1. Fusion Zone (FZ): The weld deposit contained tiny graphite nodules in an austenitic matrix, since an iron-nickel electrode was used to produce the weld.

2. Carbide Zone (CZ): This portion of the heat-affected zone is composed of partly diluted base metal that melted during the weld pass and solidified as white iron containing massive and often continuous iron carbide.

3. Partial Fusion Zone (PFZ): In this region, the portion of the matrix of the base metal near the primary graphite nodules melted during the weld pass, while the remainder of the matrix transformed to austenite. Both the liquid and the austenite were enriched in carbon by partial solution of the graphite nodules. On cooling, the liquid solidified as white iron, containing massive iron carbides, while the high carbon austenite transformed to coarse martensite.

4. Transformation Zone (TZ): This portion of the heat-affected area was heated above the critical temperature so that austenite formed during the weld pass; however, no melting occurred. The amount of carbon dissolved in the austenite depended on the peak temperature reached during the weld pass and increased towards the fusion zone. On cooling, a gradation of transformation products was obtained, ranging from coarse, high carbon martensite near the fusion zone to fine, lower carbon martensite and eventually to fine pearlite.

5. Base Metal (BM): The original structure of the base metal contained spheroidal graphite nodules in a predominantly pearlite matrix due to the high nickel content. A small ferrite ring was observed around most of the nodules. As a result of preheating and postheating, spheroidization of the base metal matrix occurred.

Width of the Heat-Affected Regions

The average widths of each of the three zones in the heat-affected area were measured using a Filar eyepiece on the microscope. Figure 3 shows the effect of preheat temperature on the width of each of the zones as well as on the total width of the heat-affected area. The carbide zone was very small except at very high preheat temperatures. At high preheat temperatures,

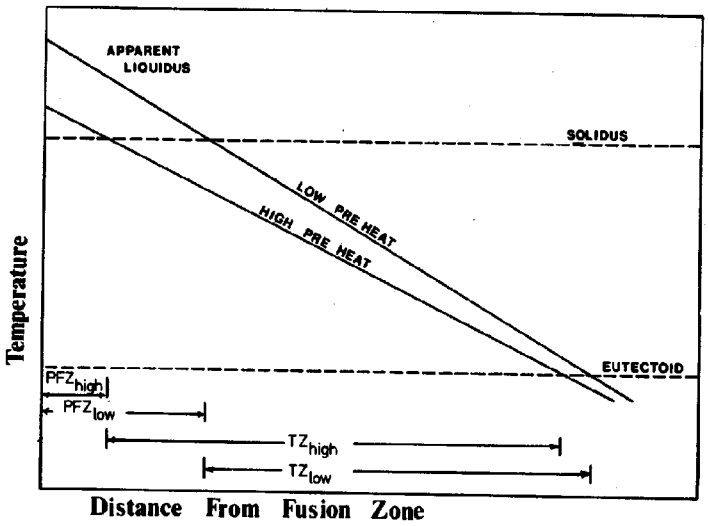


Fig. 4—A schematic diagram showing a peak temperature distribution in the heat-affected region of nodular iron weldments at two preheat temperatures

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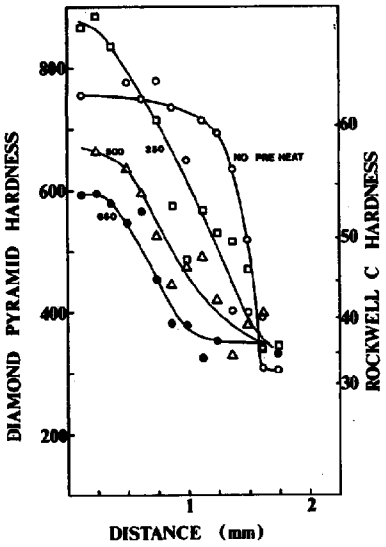


Fig. 5—The effect of preheat temperature on the hardness in the transformation zone of as-welded nodular iron weldments



Fig. 6.—Microstructure in the partial fusion zone of nodular iron weldments after welding; fusion zone is at top of photomicrograph: A—no preheat; B—930 F (500 C) preheat; C—1200 F (650 C) preheat. X250 (reduced 53% on reproduction)

the carbide region became more extensive and continuous, as has been reported previously in the available literature.[9] Carbides also precipitated more frequently at grain boundaries in the partial fusion and transformation zones at high preheat temperatures. The transformation zone increased continuously as the preheat temperature increased, as expected, while the partial fusion zone decreased in size, causing the overall width of the heat-affected area to also decrease.

The schematic diagram showing the effect of preheat temperature on the peak temperature curve (Fig. 4) may help explain these observations. Melting of the iron is not instantaneous upon heating to the liquidus temperature, since time is required for carbon to dissolve and diffuse from the primary graphite nodules and eventually combine with the iron to produce a liquid. Because of this nonequilibrium effect, the effective liquidus temperature on heating, above which melting is complete, may be higher than the equilibrium liquidus.

At high preheat temperatures, the rate of solution and liquation is faster than at low preheat temperatures, thus reducing the effective liquidus and permitting smaller fusion zones, larger transformation zones, but an overall smaller total heat-affected area. In Fig. 4, the peak temperature profile for the low preheat temperature is depicted as being above that for the higher preheat temperature; further from the weld, the profile for the low preheat temperature should fall below that for the high preheat temperature.

The actual dependence of these widths on the preheat temperature may differ depending on welding conditions, base metal thickness, and base metal microstructure. Because there was an increasing amount of ferrite in the base metal as the preheat temperature increased, and because ferritic microstructures are known to reduce transformation rates during heating, the total heat-affected zone might be expected to be smaller at high preheat temperatures.

Microhardness of the Transformation Zone

The characteristics of the formation of secondary graphite during tempering of martensite depend on the coarseness and carbon content of the martensite which is produced. One indication of the carbon content in the martensite is its hardness. Microhardness traverses were made through the transformation zone in order to compare the tendency of the martensite to temper and produce secondary graphite. The effect of preheat temperature on the diamond pyramid hardness in as-welded nodular iron is shown in Fig. 5.

When the preheat temperature is 482 F (250 C) or less, a significant amount of high carbon martensite is produced in the transformation zone. The hardness exceeds Rockwell C 60 in these regions, consistent with a large fraction of high carbon martensite produced by rapid cooling rates achieved after welding. At higher preheat temperatures, the cooling rates were slower after welding, giving less martensite, more fine pearlite, and consequently lower hardnesses in the transformation zone.

The as-welded microstructure in the partial fusion zone and in the transformation zone near the weld is shown in Fig. 6 after different preheat temperatures. The martensite is very coarse and abundant at low preheat tempera-



Fig. 7—Microstructure in the partial fusion and transformation zones before and after tempering at 1250 F (677 C), showing the effect of initial matrix structure on formation of secondary graphite: A—PFZ (as-welded); B—PFZ (tempered); C—TZ (as-welded); D—TZ (tempered). X250 (reduced 52% on reproduction)

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tures and disappears at the highest preheat temperature. Consequently, little secondary graphite caused by graphitization of the matrix would be expected during tempering after high preheat temperatures; practically no martensite is present, while secondary graphite does not form during the ferritization of pearlite.

However, greater quantities of more continuous iron carbides, including grain boundary carbides, are observed in Fig. 6 as the preheat temperature increases above 932 F (500 C). After tempering, the appearance of chain-like secondary graphite nodules from graphitization of the carbides would be expected in welds produced at high preheat temperatures.

The amount, hardness, and coarseness of the martensite also decreased with increasing distance from the fusion zone, producing a change in the incidence and morphology of secondary graphite nodules in each weld. Figure 7 shows the original as-welded microstructure and the tempered microstructure in a plate welded with no preheat, revealing the difference in matrix structure on secondary graphite formation. At distances more remote from the fusion zone, the peak temperature was lower, less carbon dissolved in the austenite, less austenite transformed to martensite, and consequently fewer secondary graphite nodules formed during tempering.

Postheat treatment reduced the hardness in the heat-affected area due to tempering of the martensite, graphitization of carbides, and ferritization of the base metal matrix. Figure 8 shows the hardness profile in the heat-affected area of a weld produced with no preheat both before and after tempering. The hardness in the transformation zone is relatively uniform after tempering and decreases as the tempering temperature increases.

Ferritization of both the heat-affected area and the base metal was slow in nodular iron due to the pearlite stabilization caused by the high nickel content of the iron. The hardness readings do not necessarily reflect the true hardness of the matrix, since it was almost impossible to completely avoid secondary graphite nodules during testing. Figure 9 shows microhardness indentations which overlap several nodules, resulting in a lower hardness value. In fact, the hardness in the base metal was often higher than in the transformation zone due to the presence of the secondary graphite.

Secondary Graphite

When martensite in nodular iron is heated, the initial stages of tempering resemble that of steel, where cementite is the eventual product. However,

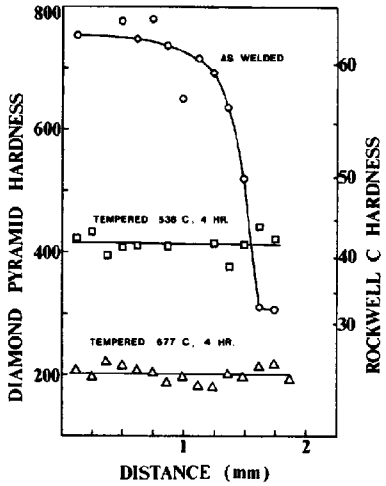


Fig. 8—The effect of postheat treatment on hardness in the transformation zone of nodular iron weldments. No preheat

cementite is redissolved and a very fine dispersion of carbon particles, or clusters, is produced during the tempering of quenched nodular iron. These carbon particles eventually coalesce and produce distinct secondary graphite nodules in a ferritic matrix on tempering at 1000 F (538 C) or above.

The number of secondary graphite nodules depends on the tempering temperature and time, on the martensite coarseness and carbon content, and on the composition of the iron. For instance, silicon increases the number of secondary graphite nodules, while copper and nickel decrease the number.[4] Similar behavior should be expected in nodular iron weldments, where martensites are produced having a wide range of previous thermal histories.

Figure 10 shows the number of secondary graphite nodules produced at three locations in the heat-affected area as a function of the preheat temperature after tempering 1 hour (h) and 4 h at 1250 F (677 C). These areas include locations near the boundaries between the carbide and partial fusion zones, between the partial fusion and transformation zones, and between the transformation zone and the base metal. The number of secondary graphite nodules decreases as the preheat temperature increases; less martensite forms in the transformation zone due to slower cooling rates. With less martensite, and consequently fewer martensite plate boundaries and intersections, the number of nucleation sites for secondary graphite are reduced.

The number of secondary graphite nodules also decreases as the distance from the fusion zone increases. Due to slower cooling rates and lower peak temperatures, the martensite contains

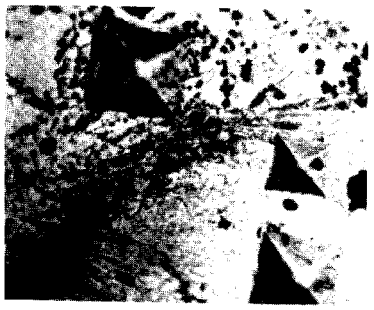


Fig. 9—Microhardness indentations overlap secondary graphite nodules, reducing hardness values for the matrix. X1000 (reduced 50% on reproduction)

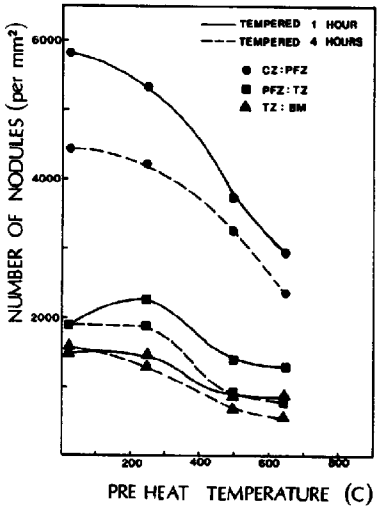


Fig. 10—The effect of preheat temperature and tempering time on the number of secondary graphite nodules produced during tempering at 1250 F (677 C) at three locations in the heat-affected region

less carbon and is less likely to form, again reducing the number of sites at which graphite can nucleate as well as reducing the amount of carbon available in the martensite to form graphite. There are some secondary graphite nodules even near the base metal; there apparently is some base line number of nodules, for which nuclei were produced during the transformation processes occurring during welding and heat treatment.

Finally, the number of secondary graphite nodules is reduced by use of longer tempering times. Some of the less favored nodules dissolve and the carbon re-precipitates on other secondary graphite particles, causing their growth. This appears to occur more rapidly when large numbers of nodules are present, as near the fusion zone.

The relationship between nodule number and nodule size is shown in

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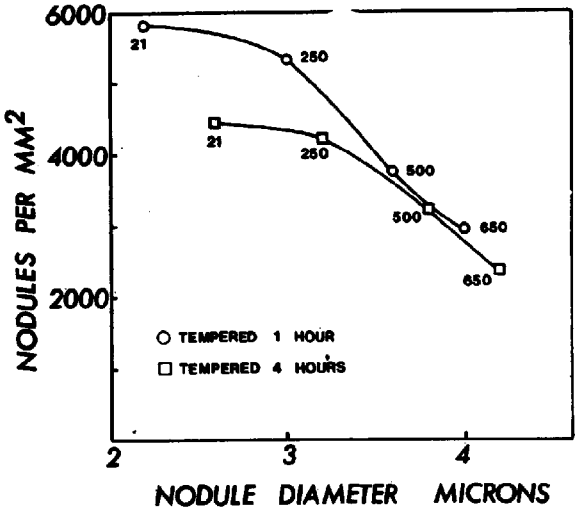


Fig. 11—Relationship between the number and size of secondary graphite nodules formed during tempering at 1250 F (677 C). Numbers refer to preheat temperature

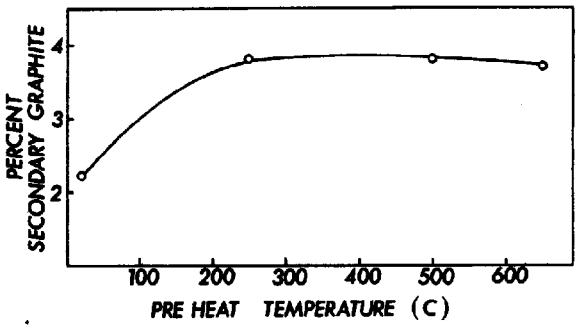


Fig. 12—The effect of preheat temperature on the amount of secondary graphite in the partial fusion zone while tempering at 1250 F (677 C) for 1 h

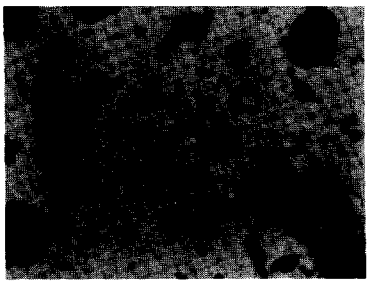


Fig. 13—Elongated secondary graphite nodules produced while tempering following high preheat temperatures. X500 (reduced 50% on reproduction)

Fig. 11. Both increasing the original preheat temperature (shown as numbers on the graph) and tempering time reduced the number of secondary graphite nodules and increased their size. Long tempering times permitted coalescence and growth of favored secondary graphite nodules, as explained previously. High preheat temperatures, on the other hand, reduced the amount of martensite, and thus the number of nucleation sites at which the nodules could form. However, due to the higher peak temperatures that would be expected at high preheat temperatures, any martensite that does eventually form should have a high carbon content and should permit large secondary graphite nodules to form. Thus, only a few, large secondary graphite nodules should precipitate when tempering the weld after high preheat temperatures were employed.

Figure 12 also shows that high preheat temperatures will cause a large amount of carbon to be made available for secondary graphite formation. The amount of secondary graphite, in area 1%, was estimated from the average number and diameter of the nodules. Although large numbers of secondary graphite nodules form at low preheat temperatures, the time at the peak temperatures is short and the martensite tends to be relatively low in carbon, since there is insufficient time to dissolve large amounts of carbon from primary graphite nodules in the austenite.

At higher preheat temperatures, more carbon is dissolved in the austenite. Although fewer nucleation sites for secondary graphite nodules are produced, there is a greater amount of carbon available to form the nodules. At very high preheat temperatures, the total percent of secondary graphite begins to decrease again, even though there is a large amount of available carbon, because there are too few nuclei. Most of the austenite transforms to pearlite rather than martensite. As the pearlite is ferritized, carbon re-precipitates on primary nodules rather than as secondary graphite nodules.

Morphology of Secondary Graphite

The shape of the secondary graphite nodules is significantly affected by the carbon content of the martensite; secondary graphite may be round and randomly distributed, as in Fig. 1, or may be elongated, as in Fig. 13. Generally, the higher carbon martensites encourage elongated secondary graphite; consequently, this morphology is found in the partial fusion zone and in the transformation zone nearest the fusion zone. Round secondary graphite is observed in the transformation zone nearer the base metal, where the peak temperature was lower and a lower carbon martensite was produced.

The ferrite grains are also elongated



Fig. 14—Chains of graphite nodules in the carbide zone formed during tempering following high preheat temperatures due to decomposition of iron carbides. X1000 (reduced 38% on reproduction)

in the regions that contain high carbon martensite. These regions produce a large number of secondary graphite, often elongated. The nodules effectively pin the ferrite grain boundaries, which are initially acicular in shape, outlining the former martensite plates, thus preventing the grains from growing into the equiaxed morphology which is more desirable. When fewer secondary graphite nodules are present, as when high preheat temperatures are used or at distances further from the fusion zone, the ferrite grains will more likely grow and become equiaxed during tempering.

Graphitization of Carbides

A second source of secondary graphite nodules is the iron carbide present in the carbide zone. The carbides form during solidification of the molten cast iron. Only very slow cooling rates can prevent their occur-

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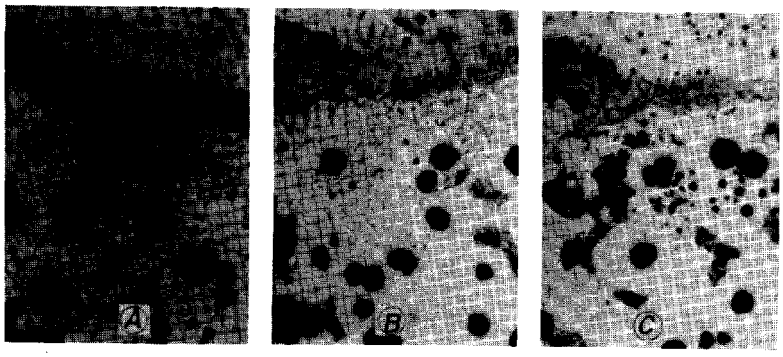


Fig. 15—Microstructure of two nodular iron weldments after annealing at 1650 F (900 C) for 2 h. Welds were preheated at three different temperatures: A—no preheat; B—480 F (250 C) preheat; C—1200 F (650 C) preheat. X250 (reduced 53% on reproduction)

rence; the typical preheat temperatures often tend to increase the amount and continuity of the carbides rather than prevent their formation.

During tempering at high temperatures, the iron carbides graphitize, with the graphite particles often preferentially nucleating at the interface between the fusion zone and the heat-affected area. Long chains of discrete secondary graphite nodules are formed, producing an almost continuous film of graphite—Fig. 14. The severity of the chain-like nodules worsens at the higher preheat temperatures, where problems with martensite and the secondary graphite from the martensite are least pronounced. Consequently, it is difficult to select a preheat temperature that will prevent the carbides, the martensite, and the secondary graphite nodules that are produced from both.

Annealing

Specimens from the nodular iron welds were also annealed at 1650 F (900 C) for 2 h and furnace cooled. Problems with secondary graphite formation were largely reduced compared to the results obtained from subcritical tempering. At intermediate preheat temperatures, very few secondary nodules were observed. A somewhat greater incidence of nodules was observed at the highest preheat temperature; these nodules probably formed from graphitization of the carbides during annealing. However, the secondary graphite nodules, although closely spaced, did not have the chain-like morphology.

The greatest number of secondary graphite nodules was observed when no preheat was used; these nodules probably formed from nuclei produced by the original martensite. Previous work has shown that these secondary graphite nuclei are not eliminated by re-austenitizing.[8] Examples of the annealed microstructures are shown in Fig. 15; complete elimination of the pearlite was not obtained due to the presence of nickel in the base metal.

Summary

In order to reduce the hardness and brittleness in the heat-affected area of nodular iron weldments, the formation of martensite and carbides must be prevented or these constituents must be eliminated by heat treatment. However, any heat treatment must be carefully controlled to minimize secondary graphite nodules and to maximize the ductility in the weld.

Increasing the preheat temperature produces a number of desirable effects. First, high preheat temperatures, by reducing the cooling rates, minimize the formation of martensite in the transformation zone. In addition, the size of the partial fusion zone, which contains high carbon martensite as well as carbides, is reduced. Finally, due to the formation of coarse martensite, relatively few secondary graphite particles will grow during tempering; in fact tempering may not be necessary if the preheat was sufficient to reduce the amount and continuity of the martensite.

Unfortunately, high preheat temperatures may increase the size and continuity of the carbide zone and the carbon content in the martensite. Unless the carbides are rendered discontinuous by proper control of the welding parameters and the base metal microstructure, postheat treatment will be necessary and elongated secondary graphite and chains of secondary graphite may be produced if a subcritical postheat temperature is used.

Postheating is necessary in most cases in order to eliminate the carbides and martensite and thus reduce hardness and brittleness. Although low tempering temperatures can reduce the hardness of martensite without causing secondary graphite formation, higher postheats are required to graphitize the carbides; consequently, secondary graphite nodules will form. The number of secondary graphite nodules will diminish slightly with increasing postheat times.

Fully annealing the weld appears to most effectively minimize the secondary graphite nodules; however, it is not successful in completely preventing their formation. The most effective way to prevent or minimize secondary graphite nodules is to prevent the formation of carbides and martensite during welding. This requires careful control of the base metal composition and microstructure, the preheat temperature, and other welding parameters.

Acknowledgments

The authors gratefully acknowledge the generous assistance of the Welding Research Council for its partial support of this research and of Wagner Casting Company for providing chemical analysis.

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Short-arc Welding of S.-g. Iron in the SKF Katrineholm Works, Sweden

By S. A. Fröberg*

The author discusses, in a practical manner, the application of short-arc welding to the correction of machining errors, repair of defects, and the joining of castings, made from s.-g. iron, to other metals. Details are given of physical, mechanical and micrographic investigations of such welds.

In the SKF Katrineholm works, the short-arc technique has been used to weld s.-g. iron for almost one-and-a-half years, particularly for, **1** correcting machining errors, **2** repairing defects revealed by machining, and **3** welding liners in austenitic manganese-steel, or hardened spring-steel to s.-g. iron axle-boxes. The author's experience suggests that it is the most satisfactory method of welding s.-g. iron.

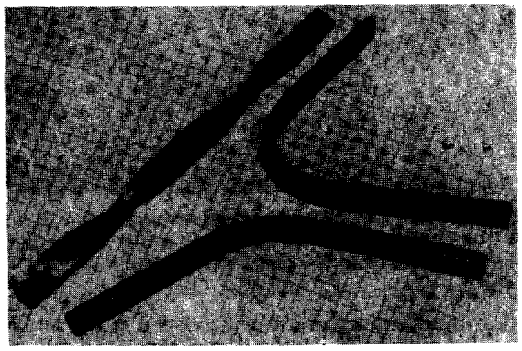
In the short-arc technique, welding conditions are adjusted to produce a short-circuiting arc in which the weld metal is transferred in the form of large droplets. Since short circuiting reduces the energy generated and, thus, also the temperature of the molten metal, melting and heating of the base metal are minimized. It is important that the welding electrode should be capable of holding carbon in solution and should be of small diameter (Nickel 61 filler wire, 0.03 in. (0.75 mm) in diameter, has been found to be the most satisfactory electrode in this respect) and, for best results, the electrode should be given a swiftly-oscillating transverse movement as it is moved forward and, to decrease heat input further, the weld should be produced in a series of short lengths. Due to the low energy which characterizes the process, finish-machined parts may be welded without risk of excessive undercutting. Table 1 lists typical welding data.

*Mr. Fröberg is attached to SKF Sweden.

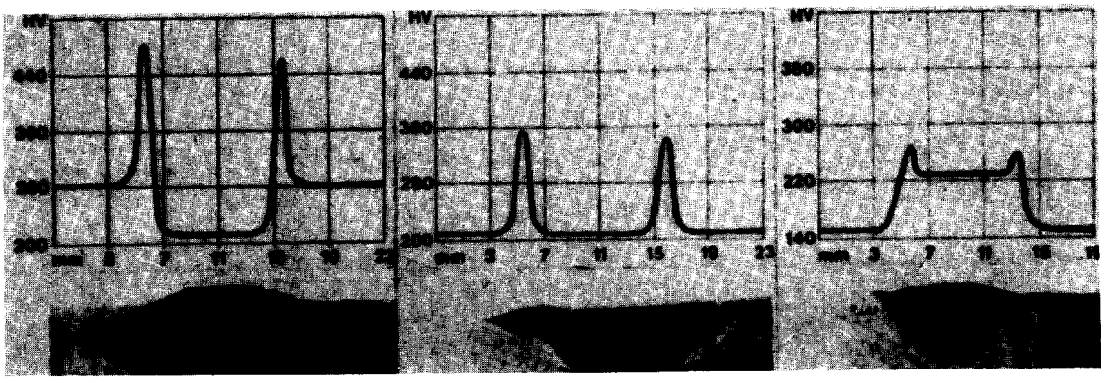
Test Welds

To determine the variation in structure and hardness in the weld zones, test welds were made in a large number of 12 by 75 by 150 mm plates in SKF's normal grades of s.-g. iron—the ferritic iron SKF G83, the pearlitic/ferritic iron G82 and the pearlitic iron G81. A 6-mm-deep V-joint with a 75 deg. included angle was machined along the longitudinal centre line of the plates, and welding was effected using both single and multiple beads. Optimum welding conditions are presented in Table 2, and data on the hardness, mechanical properties and structure of the welds are tabulated in Table 3. For purposes of discussion, the heat-affected zone has been divided into three arbitrary zones: **1** The melted zone; **2** the conversion zone, in which the parent metal has been heated above the critical temperature, and **3** the modified-structure zone—that where the metal has been heated to below the critical temperature but where some modification has taken place.

1



1 Torsion- and bend-tested specimens in ferritic iron.
2 Hardness variation across, and width of, the weldment in pearlitic iron.
3 Hardness variation across, and width of, the weldment in pearlitic/ferritic iron.
4 Hardness variation across, and width of, the weldment in ferritic iron.

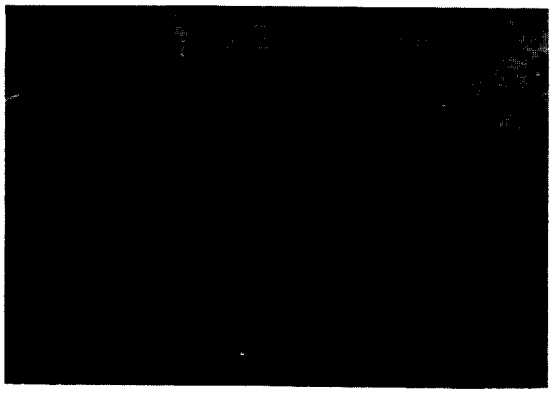


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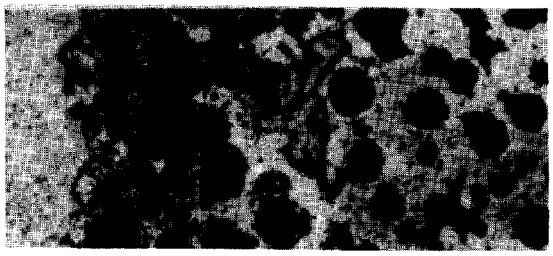
Source: Foundry Trade Journal, May 23, 1968

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6



7



5 *Microstructure of short-arc weld in ferritic iron.*
6 *Microstructure of metal-arc weld in ferritic iron.*
7 *Microstructure of short-arc weld in pearlitic/ferritic iron.*

Tensile Tests

Cast tensile-specimens were sectioned in the middle, welded with an X joint and then tested. Usually the test-bars in G83 broke in the parent metal, and the stronger bars in the G82 and G81 grades fractured in the welds.

TABLE 1.—*Welding data.*

Welding machine:	ESAB A9 PP/25
Welding electrode:	Nickel filler metal 61; dia., 0.03 in. (0.75 mm.)
Arc voltage:	16–20 V.
Welding amperage:	90–120 A.
Protective gas:	Argon, 6–8 litres per minute.
Wire feed:	6–8 m. per minute.

TABLE 2.—*Optimum welding conditions.*

Arc voltage:	17 V.
Welding amperage:	100 A.
Welding wire:	Nickel filler metal 61; dia., 0.03 in. (0.75 mm).
Protective gas:	6 litres per minute.
Wire feed:	6 m. per minute.

Torsion and Bend Tests

To ascertain the ductility of short-arc-welded s.-g. iron, particularly the ferritic grade, 10 by 25-mm rectangular-section or 25-mm-dia. round-section welded test-pieces were twisted or bent. Fig. 1 shows specimens after testing.

Hardness Tests

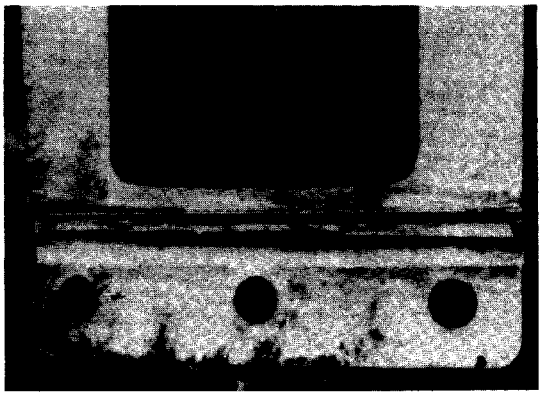
Hardness across the heat-affected zone varied with the grade of s.-g. iron. In the case of the pearlitic grade, the maximum increase in hardness in the heat-affected zone was about 200 Brinell and the width

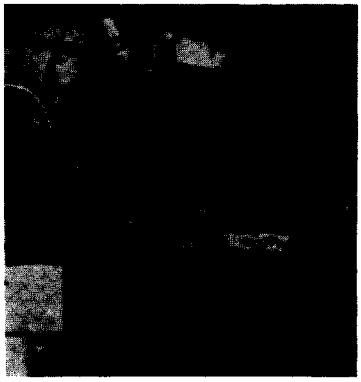
TABLE 3.—*Weld properties and depth and structure of the welding zone.*

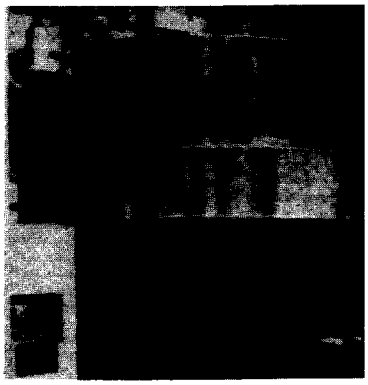
Grade SKF Swedish Standard (SIS)	G81 0707	G82 0727	G83 0717
Properties			
Proof Stress, kg. per sq. mm.	40	35	28
Tensile strength, kg. per sq. mm.	60	52	40
Elongation, per cent	2	7	15
Brinell hardness	300 max.	170–230	140–190
Matrix	Pearlitic	Pearlitic-ferritic	Ferritic
Depth of the welding zones, mm., approx.			
Melting zone	0.0	0.0	0.0
Transition zone	0.3	0.2	0.1
Modified-structure zone	0.4	0.2	0.0
Total depth of the welding zone	0.7	0.4	0.1
Structure of the welding zones			
Melting zone	Traces of primary cementite	Traces of Primary cementite	Insignificant traces of primary cementite
Transition zone	Predominantly martensite	Predominantly martensite	Predominantly martensite
Modified-structure zone	Heat-affected pearlite	Heat-affected ferrite	—

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of the zone was 0.7 mm (see fig. 2); the corresponding values for pearlitic/ferritic s.-g. iron were 140 Brinell and 0.4 mm (fig. 3). The lowest values were exhibited by ferritic s.-g. iron, the hardness increase being not more than 110 Brinell and the width of the heat-affected zone only 0.07 mm: In this case, the maximum hardness was no more than about 260 Brinell, i.e. only about 30 Brinell higher than that of the weld metal (see fig. 4).

The short-arc technique can be employed to greatest advantage in welding ferritic s.-g. iron, the width of the martensitic portion of the heat-affected zone being only 1/100 that associated with ordinary arc-welding (see figs. 5 and 6).

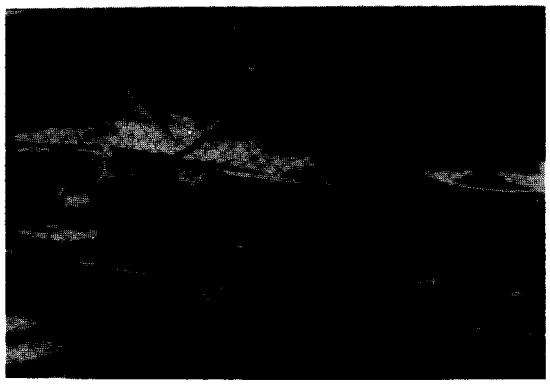
Microstructure

Figs. 5 and 7 illustrate the microstructure of the heat-affected zones, and show clearly how much narrower the zone is in the ferritic iron. The spheroids adjacent to the weld are, to a certain extent, surrounded by cementite, but the effect of the latter is believed to be relatively insignificant.

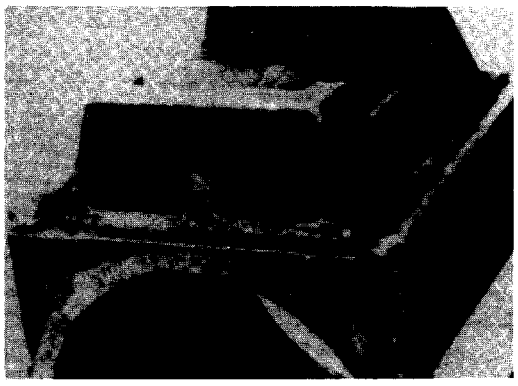
Applications

Fig. 8 shows a 522-kg SKF G82 bearing housing

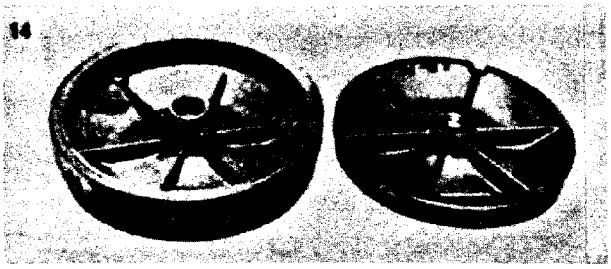
12 ***Detail of repair-welded roller-bearing axle-box.***



13 ***Roller-bearing axle-box with welded-on liner.***



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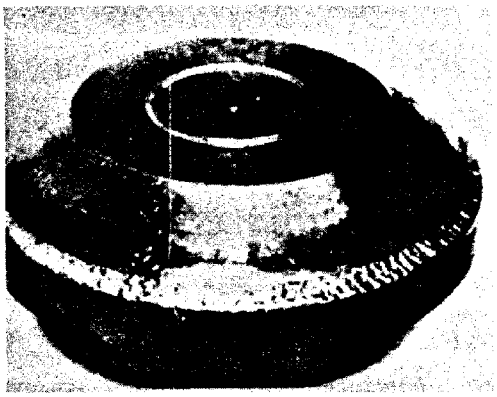
14 *Components of piston before welding.*
15 *Piston after welding.*
16 *S.-g. iron detail welded to a steel plunger.*

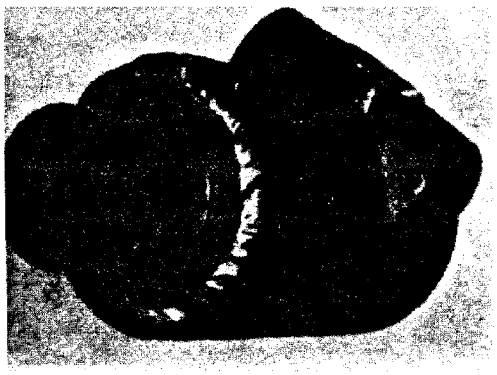
for a paper machine. A fault introduced during machining of the groove in the base of the housing was corrected by short-arc welding. Fig. 9 is a photograph of part of the base after the weld had been ground.

The sleeve illustrated in fig. 10 is in SKF G82 iron and weighs 62 kg. Machining revealed slight but extensive casting defects, which were subsequently remedied by short-arc welding. Fig. 11 shows the sleeve after final grinding of the weld. The two roller-bearing axle boxes in SKF G82 iron pictured in fig. 12 each weigh 49 kg. In the box on the right a core was placed the wrong way round, causing the boss for a bolt hole to be cast in the wrong place, and a new boss had to be built up by short-arc welding. In the box on the left the boss is in the correct place. Fig. 13 shows a 25-kg roller-bearing axle box in SKF G82 iron with welded-on liners in austenitic manganese-steel.

Fabricating a Piston

In figs. 14 and 15 the two components of a piston can be seen before and after fabrication welding. The piston, produced in pearlitic ferritic s.-g. iron, is designed to operate under a pressure of 6 atmospheres. Components such as those shown in fig. 16—an s.-g. iron detail welded to a steel plunger—have so far given more than a year's trouble-free service.





Conclusion

Though the author has approached short-arc welding with caution, his experience to date has been very encouraging. SKF's Swedish customers have shown keen interest in welding s.-g. iron, and it is believed that the short-arc process will go far to solve the problem of welding graphitic materials. The writer considers it the best method so far available and believes that it offers promises of widening still more the field of application of s.-g. iron.

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Welding Ductile Iron using the GMA Short Circuiting Transfer Process—A Progress Report

Although ductile iron production has established an enviable growth pattern further growth has been impeded by the difficulty in obtaining good welds.

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Madison

Abstract

An apparent lack of weldability in ductile iron, generally attributable to three principal factors, is investigated here, using previous studies as a point of departure. Fully automatic short circuiting arc deposition of nickel on ductile permits employment of much larger diameter filler wire than used previously while still obtaining noncontinuous fusion line carbides. This larger diameter wire is less costly on a per pound basis and permits higher deposition rates. The present study is being continued.

Introduction

ALTHOUGH DUCTILE IRON PRODUCTION has established a growth pattern which exceeded even the most optimistic predictions, the further utilization of this material has been impeded somewhat by the present difficulty in obtaining ductile and machinable welds. This apparent lack of weldability is generally attributed to three factors:

1) The formation of martensite in the heat affected zone due to the heating of the base metal into the region of austenite formation, followed by the rapid cooling characteristic of most fusion welding processes.

2) The formation of iron carbide in the junction zone due to the heating of the base metal into the temperature region between the liquidus and solidus, followed by rapid cooling incapable of permitting graphite formation.

3) The formation of martensite and/or iron carbide in the weld metal due to the rapid cooling of the weld metal (dependent upon the final weld metal chemistry).

The formation of either martensite or iron carbide in the weld metal or base metal is undesirable if ductile iron is to be successfully welded. Iron carbide, formed during solidification, and martensite, formed during cooling below the lower critical temperature range, both contribute to a hard, brittle condition at the weld. Accordingly, the structure shows very low ductility and will be prone to cracking during welding. This cracking susceptibility is induced by the deformation required to relieve stresses imposed by the uneven heating and cooling cycles of the welding process.

Martensite formation can be eliminated by preheating the base plate to temperatures such that the cooling rate is reduced sufficiently to avoid the martensite transformation, approximately 600 F (316 C). If the base plate is not preheated, the formation of martensite can be controlled somewhat by controlling the welding process, causing the base metal to cool at a slow rate wherever it was heated into the austenite formation temperature region. Further control of martensite formation in the weld metal can be obtained by using a filler metal composition incapable of martensite formation even when diluted with iron and carbon from the base metal.

Carbide formation presents a more formidable barrier to the welding of ductile iron (Fig. 1). Very unusual welding procedures or extensive preheating to temperatures greater than 1500 F (816 C)[1] are required to slow the solidification cooling rate to a level such that iron carbide does not occur in the junction zone. Welding at such high preheating temperatures is uncomfortable and results in undesirable scaling and distortion of the casting. Post-heating to remove carbides from the weld area also requires

Source: AFS Transactions, 67-94, 1967 (American Foundrymen's Society)

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Fig. 1. Carbidic network at junction zone of ferritic ductile iron weld obtained by melting through a nickel layer with 0.035 in. dia. mild steel at 180 amps, 34 volts, 16 ipm torch travel speed (23,000 joules per in.). The carbidic network is typical of conventional arc welds. Nital etch, 100×.

the use of high temperatures over 1650 F (899 C) for long periods of time and controlled cooling rates.[2]

Conventional welding of ductile iron requires that the process be capable of producing a welded joint free of cracking and having sufficient strength and ductility to perform adequately in service. In order to obtain such a weld the use of gas welding, at very low travel speeds, is advocated. This permits a controlled slow cooling from the liquid and thereby minimizes the formation of carbides during solidification. If the structure is to be machinable, full annealing of the weld and its associated heat affected zone may be required to remove all traces of continuous carbide formation.

Since electric arc welding is usually accompanied by greater temperature gradients between the fusion zone and the base metal, the formation of carbides is more prevalent. Welding may be carried out satisfactorily if the weld is produced without cracks. This requires that the weld be placed only in relatively unrestrained geometries. Following welding, the weld and its heat affected zone must then be fully annealed to remove the continuous carbides which have formed. It should be noted, however, that if the carbides found in either the weld metal or the heat affected zone are discontinuous, that is, discrete isolated patches of the ledeburitic eutectic, only a modest influence on machinability or mechanical properties will be obtained.

More recent developments in welding have resulted in the GMA (gas metal arc) short circuiting transfer process capable of being used as a fully automatic or semi-automatic process. Briefly, this process utilizes a constant potential power source with "slope" control (which limits the peak current when a short circuit exists), a relatively small diameter continuous filler wire and a shielding gas appropriate for the filler wire and base metal. The wire feed rate into the arc is machine controlled and the torch travel speed may be manually or machine controlled. Fusion of the base metal and wire is accomplished by a series of high-current, low-voltage "shorts" (20 to 200 shorts per second) wherein the wire dips into the weld pool, extinguishing the arc. During each short, enough wire melts to reestablish the arc, permitting the process to repeat itself. The heat input from this process can be approximated from the simple formula:*

$$Q = 60\,EI/S$$

where: Q = heat input, joules per in.
E = average arc voltage, volts
I = average arc current, amperes
S = welding torch travel speed, in. per minute.

The short circuiting arc process permits the consideration of another approach to the welding of ductile iron. During welding, melting starts at the graphite spheroid-austenite interface and proceeds as carbon from the spheroid diffuses through the liquid and into the austenite. This diffusion of carbon is time-dependent—that is, the extent of melting in the junction zone can be minimized by shortening the time of exposure of the junction zone to temperatures above the solidus. The short circuiting transfer GMA process provides a method of welding with low heat input and thus offers the possibility of welding ductile iron with a fusion zone containing minimum iron carbide (or, at most, discontinuous iron carbide) and not seriously impairing the ductility and machinability of the joint.

Some study of the use of the semi-automatic short circuiting arc for ductile iron welding has been reported in the literature.[3,4] These studies have considered the use of fine diameter wires (0.030 in.) bearing large percentages of nickel (57-97%). Other work also considered the use of mild steel wires of the same size; however, severe cracking of the weld metal occurred.

This study was undertaken to further investigate the application of the fully automatic short circuiting arc to the welding of ductile iron, with specific emphasis on the use of larger diameter filler wires capable of yielding higher deposition rates and lower weld costs.

Results and Discussion

Preliminary Welds, Mild Steel Filler Wire—Ductile iron from a number of commercial foundries was used in this study. Castings were prepared as plates 5 x 6 x 1 in. and were of a nominal 4.45% carbon equivalent composition. Single or double vee butt joints were prepared with a shaper after the plates were sectioned lengthwise, Fig. 2.

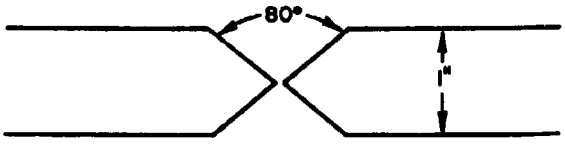


Fig. 2. Sketch of typical double vee butt joint used in this study.

*This formula is an approximation because it ignores the effect of power factor and the loss of heat by radiation and convection from the arc.

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The first five samples were prepared using the following welding conditions:

Wire diameter	0.35 in.
Wire feed speed	193 ipm
Contact tube height above base metal	0.5 in.
CO_2 gas flow rate	30 fh
Torch travel speed	32.4 ipm
Average arc current	100 amps
Average arc voltage	19 volts
Heat input	3520 joules/in.
Root gap	0.3 in.

This series of welds was prepared to determine the effect of joint geometry on the formation of carbides within the weld. No preheat or post heat was employed and the welds were sectioned transversely on an abrasive cutoff wheel and examined macro- and microscopically. Lack of fusion was apparent in all welds; however, only discontinuous carbides were present along the entire junction zone of all welds. Martensite and retained austenite were present on both sides of the fusion line. Cracking was considerable in the martensitic zones, Fig. 3.

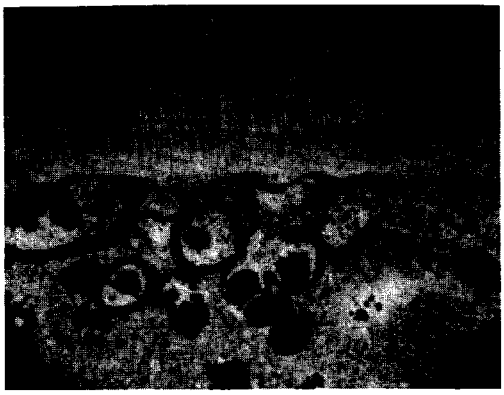


Fig. 3. As-welded junction zone of steel deposit on pearlitic ductile iron at 19 volts, 100 amps, 32.4 ipm torch speed (3520 joules per in.). Note the discontinuous carbides and cracks. An abrasive cutoff wheel was used in preparation of this specimen. Nital etch, 100×.

Since an abrasive cutoff wheel was used to prepare the specimens in this series, it was not possible to attribute the cracking specifically to either the cutoff operation or the stresses present during welding. Therefore, another sample was prepared (cut with a hacksaw) and examined after tempering the weld for two hours at 900 F (482 C). The lack of cracking as shown by Fig. 4 indicates that the cracking shown in Fig. 3 could be primarily attributed to thermal stresses due to the abrasive cutoff operation.

All remaining mild steel welds were therefore prepared using an 80 deg. double vee joint, but were tempered for two hours at 900 F (482 C) prior to micro- or macrostructural examination. In order to obtain more complete fusion of the weld metal, the welding heat input was increased by reducing the torch travel speed for those weld passes directly



Fig. 4. Junction zone of steel deposit on pearlitic ductile iron at 19 volts, 100 amps, 32.4 ipm torch travel speed (3520 joules per in.) after two hours post heat at 900 F. The joint is the same as in Fig. 3. Note the discontinuous carbides and lack of cracking. Nital etch, 100×.



Fig. 5. Junction zone of steel deposit on pearlitic ductile iron at 19 volts, 100 amps, 22.7 ipm torch travel speed (5020 joules per in.) after two hours post heat at 900 F. Note that the carbides are larger than those of Fig. 4. Nital etch, 100×.

contacting the ductile iron. Passes not in contact with ductile iron were made using increased welding current as well.

At a heat input of 5020 joules per in. (22.7 ipm), carbides in the junction zone were discontinuous but larger than in the previous samples (3520 joules per in.), Fig. 5. At still higher heat input, 7430 joules per in. (16.1 ipm), carbides were still larger and tended to form network structures connecting the graphite spheroids at the junction zone, Fig. 6. It was determined that a heat input of approximately 5000 joules per in. should be further studied from the point of view of joint ductility.

Evaluation of joint ductility was performed by using the standard 180 deg. side bend test. This test uses a transverse weld specimen machined to ⅜ in. thickness and bent around a 1½ in. diameter pin. Since the pearlitic ductile iron did not pass this test even in the unwelded condition, all subsequent welding was performed on ductile iron given a

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Fig. 6. Junction zone of steel deposit on pearlitic ductile iron at 19 volts, 105 amps, 16.1 ipm torch travel speed (7430 joules per in.) after two hours post heat at 900 F. Note the carbide networks between graphite spheroids. Nital etch, 100×.

ferritizing anneal (less than 10% pearlite).

Thus, a ferritic ductile iron specimen was then welded using the following parameters. It may be noted that a globular arc transfer technique was used for those passes not in contact with ductile iron, Fig. 7, made using the following welding conditions:

	Short Circuiting Arc	Globular Transfer
Current	110 amps	170 amps
Voltage	19 volts	25 volts
Torch travel speed	24.4 ipm	24.4 ipm
Heat input	5140 joules/in.	10,500 joules/in.

Side bend specimens of these welds fractured after bending less than 20 deg. Failure began at the corner of the fusion line and progressed through the base metal, Fig. 8. No yielding was observed in the weld metal; however, base metal yielding prior to fracture was readily apparent.

Tensile specimens failed at 58,000 to 61,700 psi with fractures similar to those obtained in bend tests but with a slightly greater tendency for the fracture to follow the fusion line, Fig. 9. Again, yielding was observed in the ductile iron but not in the weld metal.

Hardness measurements across the weld showed the following average results:

Base metal, ferritic ductile iron	128 Bhn*
Weld metal, tempered martensite	353 Bhn**
Weld metal, mild steel	206 Bhn**

The high hardness of the tempered martensite may explain the failure pattern of the tensile and bend specimens since yielding of the ductile iron base metal causes a notch to form at the fusion line resulting in premature failure of the joint.

*Taken as Rockwell B and converted to Brinell hardness number according to *Metals Handbook*, Vol II, *American Society for Metals*, 1964, p. 670-672.

**Taken as Rockwell C and converted to Bhn according to *Metals Handbook*.



Fig. 7. Sketch of steel point made with short circuiting and globular or spray transfer.

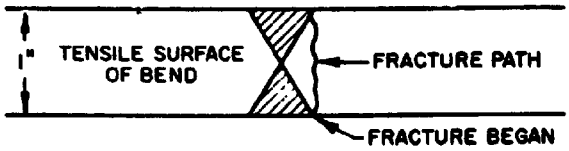


Fig. 8. Sketch of fracture pattern in side bends of ferritic ductile iron welded with mild steel at 5140 joules per in.

Nickel Alloy Filler Wire—In order to eliminate the tempered martensite layer in the weld metal at the fusion line (formed due to the high carbon and silicon content in the weld metal in those regions), nickel alloy wire was used for the short circuiting arc deposits on the base metal. Ninety-six per cent nickel wire was available in 0.062 in. diameter. This larger wire diameter (vs. 0.030 or 0.035 in.) yields higher deposition rates but requires higher operating currents. This in turn requires higher torch travel speeds to limit carbide size. Furthermore, the deposition of nickel is best achieved using an argon shielding gas rather than carbon dioxide. Welding conditions established were as follows:

Argon gas flow	30 cfh
Wire feed speed	165 ipm
Average arc voltage	18-19 volts
Average arc current	290-300 amperes
Torch travel speed	50 ipm
Average heat input	6260-6840 joules/inch

A number of specimens were prepared by "buttering" the ductile iron with nickel using the short circuiting transfer process. The faces of the ductile iron were buttered separately and then joined by the spray arc CO_2 process using 0.035 in. diameter steel wire at 180 amps, 34 volts, 16 ipm torch travel speed (22,800 joules per in.). This technique resulted in lack of root penetration by the steel, causing a void at the weld center, and "burn through" at the weld face corners due to the high heat input of the spray arc process, Fig. 10. No noticeable difference was observed between welds made using interpass cooling to room temperature and those made with consecutive nickel passes. Accordingly, all additional nickel welds were prepared without interpass cooling.

The mechanical problem of inadequate root penetration was by-passed by using a single vee, 90 deg. joint with feather edge. A copper backup strip was used to prevent burn-through of the feather edge, and "around-the-corner" nickel deposits were added at the face of the vee, Fig. 11. In all, 25 nickel passes (6480 joules per in.) and 36 mild steel passes (19,100 joules per in.) were required to complete the joint. No restraint was imposed on this plate during welding.

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Although all carbides were discontinuous, the amount of carbides present was noticeably greater at the root of the weld than elsewhere due to the greater heat concentration afforded by the joint geometry. Martensite formation was limited to a thin layer along the fusion line and around the carbides, and no cracks were found, Fig. 12.

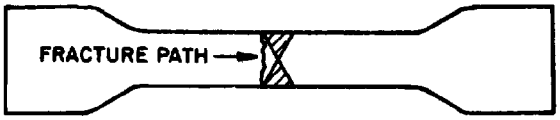


Fig. 9. Sketch of fracture pattern in tensile test specimens of ferritic ductile iron welded with mild steel at 5140 joules per in.

Severe distortion of the specimen necessitated the use of smaller test bars, 0.370 x 0.673 x 4.5 inches, with no reduced section machined on the tensile bar. This bar failed at 68,300 psi at the gripping jaw, ¾ in. from the weld. Yielding was apparent in the ductile iron and in the nickel layers but not in the steel which had cracked considerably prior to fracture of the specimen. Side bend test bars (duplicates) failed entirely in the steel deposit after bends of less than 20 deg. Some porosity was observed in the steel deposit.

Following a number of attempts to produce satisfactory welds free of mechanical lack of fusion and weld porosity, several samples were prepared using the 80 deg. double vee joint preparation. The most desirable technique used no root gap. In addition, the use of a CO_2 shielded spray arch technique for completing the joint with mild steel was discontinued and replaced with a manual stick electrode process. This process, using ⅛ in. E-7018 electrodes, is much less penetrating, and is more versatile than the previously used process. Welding conditions for these welds were as follows:

Nickel Deposits	
Wire	0.062 in. NS-61
Wire feed speed	165 ipm
Average arc voltage	18 volts
Average arc current	295 amps (DCRP)
Torch travel speed	51 ipm
Heat input	6240 joules/in.
Argon gas flow	30 cfh
Steel Deposits	
Electrode	⅛ in. E 7018
Average arc voltage	22 volts
Average arc current	120 amps
Average arc travel speed*	14 ipm
Average heat input*	11,350 joules/in.

*As the steel was deposited manually, the travel speed could not be held constant.



Fig. 10. Sketch of butt weld made by buttering the joint faces of the ductile iron separately with nickel prior to joining with steel, showing areas of lack of fusion and heavy carbides.

During steel deposition, the interpass plate temperature was not permitted to exceed 575 F (302 C). After completion of the weld, the specimens were post heated at 900 F (482 C) for two hours. Discontinuous carbide along the fusion line and a thin tempered martensitic layer in the ductile iron adjacent to the fusion line were observed, similar to Fig. 12. Side bend specimens failed after bends of 55 and 60 deg. Failure initiated in the steel and continued through both the weld deposit and the ductile iron. While the ductile iron and nickel deposits showed yielding, the steel deposits did not yield significantly.

The difference in properties across the joint is illustrated by hardness data ranges as follows:

Ferritic ductile iron base metal	122-124 Bhn*
NS-61 deposits on base metal	176-198 Bhn*
E 7018 deposit on NS-61 deposit	312-379 Bhn**

Continued Weld Formulation—The E7018 electrode is well known for the ductility of its deposited weld metal in conventional steel welding applications. However, in this application the deposit is made on a nickel base, permitting the nickel to alloy with the mild steel in an amount controlled by the amount of penetration. It may therefore be proposed that the ductility of the composite joint is limited by the solid solution strengthening of the mild steel by nickel.

A number of alternatives may thus be considered to obtain greater joint ductility:

1) Use the indicated welding procedure outlined above for nickel deposition on the base metal, followed

*Taken as Rockwell B and converted to Bhn according to *Metals Handbook.*

**Taken as Rockwell D and converted to Bhn according to *Metals Handbook.*

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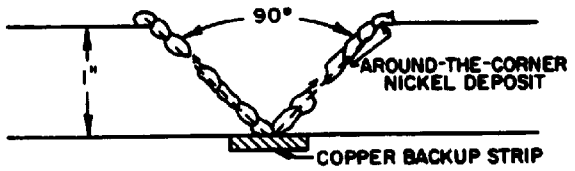


Fig. 11. Sketch of 90 deg. single vee joint made with nickel and steel showing geometry, location of nickel passes and location of around-the-clock corner passes.

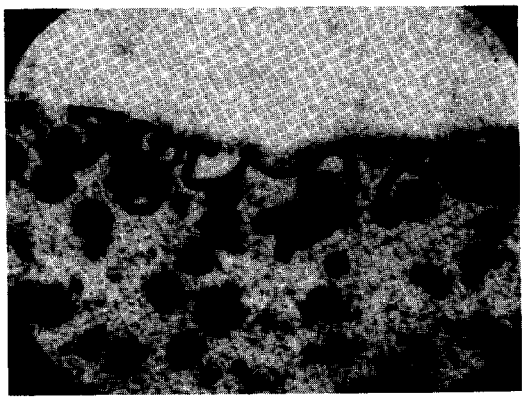


Fig. 12. Typical junction zone of nickel-steel composite butt weld in ferritic ductile iron. The 0.062 in. nickel wire was deposited on the ductile iron at 18 volts, 295 amps, 50 ipm torch travel speed (6380 joules per in.). After completion, the joint was post heated at 900 F for two hours. Note the discontinuous carbides surrounded by tempered martensite. Nital etch, 100×.

by the completion of the joint with nickel wire.

2) Use nickel alloyed steel for completion of the joint after nickel deposition to prevent excessive hardening by permitting austenite formation—e.g., austenite stainless steel over nickel.

3) Use filler metals other than nickel or devise complex filler metals which eliminate the disadvantages of the nickel filler.

Conclusions

Considering the previous studies as a point of departure, several conclusions may be drawn from the present study.

Due to the greater control of welding parameters available with fully automatic (vs. semi-automatic) short circuiting arc deposition of nickel on ductile iron, it is now possible to employ much larger diameter filler wire than previously used and still obtain non-continuous fusion line carbides. The larger diameter wire is less costly on a per pound basis and permits higher deposition rates.

A direct comparison of the ductility of joints made by the methods outlined above with that of other studies is not possible due to differences in approach to testing procedures. It might, however, be pointed out that in a recent study[3] using 0.030 in. nickel filler wire similar in composition to that of the present study, face bend angles of 35 deg as welded and 90 deg after a two hour post heat at 1112 F (600 C) were obtained with an all-nickel joint, while side bends of the nickel-steel composite joint of this study bent 55 and 60 deg. after a two hour post heat at 900 F (482 C)

A comparison between Fig. 12 and Fig. 4, 5 and 6 seems to indicate that heat input as defined previously is not the only factor limiting carbide size and continuity. The carbides of Fig. 12 (NS-61 deposited at 6240 joules per in.) are most similar in size and distribution to those of Fig. 4 (steel deposited at 320 joules per in.). Thus, there appears to be an effect of torch travel speed in addition to its role in the heat input equation and/or an effect due to differences in the filler metals. In other words, equivalent heat inputs do not necessarily result in equivalent metallurgical structures.

A further point which may be considered is that previous studies[3,4] encountered severe cracking with mild steel deposits on ductile iron with 0.030 in. wire at 130 amps and 24 volts. On the other hand, in the present study, with 0.035 in. wire at 100 amps and 19 volts, cracking is greatly reduced (Fig. 4) but lack of fusion becomes a problem. Both of these facts are probably attributable to the reduced penetration accompanying the low current and voltage employed in this study. Then by exploring welding conditions between these two extremes, steel joints may also become feasible.

The present study is being continued by considering other filler metals and techniques as outlined above. Also to be considered is the use of preheat as an alternative to post heat for dealing with martensite and the effect of preheat on carbide size and distribution.

Acknowledgment

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Preheat Effects on Gas Metal-Arc Welded Ductile Cast Iron

Preheat adversely affects fusion line carbides when using nickel filler metal in the short-circuiting arc welding of ASTM grade 60-45-15 ductile iron

BY D. J. KOTECKI, N. R. BRATON AND C. R. LOPER, JR.

ABSTRACT. This study explores the relationship of preheat to the formation of martensite and carbides when welding ASTM grade 60-45-15 ductile iron. The welding is performed by the short-circuiting transfer technique of the gas metal-arc process.

It is found that a preheat temperature of approximately 600° F is adequate to prevent martensite formation in the heat-affected zone of these weldments. However, this preheat adversely affects carbide geometry. At moderate levels of welding heat input, 5000 joules/in., carbidic areas in the partial fusion zone become nearly continuous. If preheat is not used with this same welding heat input, the carbidic areas are clearly discontinuous.

At a lower level of welding heat input, such as 1850 joules/in., discontinuous carbidic areas can be obtained with this preheat temperature. However, it appears that better bend test results, when using nickel filler metal, can be obtained with neither preheat nor postheat than with just preheat.

Bend test results and fusion line microstructures are presented for comparison.

Introduction

In the selection of a material for an engineering application, the mechanical properties such as strength, ductility, resistance to brittle fracture and machineability are a primary consideration. Furthermore, if the material is to be welded, an ideal joint in that material should have properties similar to the base metal. However, because a welded joint is usually produced under markedly different conditions from those under which the base metal was produced, this ideal is sometimes extremely difficult and costly to achieve. Such is the case for ductile iron.

Since its first announcement in 1948, ductile iron has developed into a desirable engineering material of widespread use due to its wide range of possible mechanical properties and excellent castability. Greater utilization of this material, however, has been impeded by difficulty in obtaining ductile and machineable welds. This difficulty stems from the same high carbon equivalent content which gives the material its wide range of possible mechanical properties and excellent castability. Carbon, during a fusion welding thermal cycle, tends to result in two effects:

1. Wherever carbon-rich liquid (eutectic) is formed, in the weld proper and in the zone of partial melting adjacent to the weld proper, this liquid tends to freeze as carbide because of the rapid cooling rate in the weld zone.
2. Wherever austenite is formed, in the weld proper, in the zone of partial melting and in the heat-affected zone above the lower critical temperature range, this austenite tends to transform to martensite upon cooling due to the inherent hardenability of these compositions.

Both carbide and martensite, being hard and brittle, are detrimental to the ductility and machineability of the joint, and may cause cracking during cooling.

These tendencies are always present when ductile iron is fusion welded. However, the extent to which carbide and martensite actually form is a function of local chemistry and local thermal cycle. It is by manipulating these two variables that some progress has been made towards obtaining an improved welded joint.

Both carbide and martensite formation in the weld proper can be prevented by using a filler metal capable of holding carbon in solid solution in a ductile matrix even when it is rapidly cooled. The most successful and widely used of these filler materials are nickel-based alloys (c.f., Thorneycroft[1] or Osman and Stephenson[2]).

However, the use of nickel-based filler metal, or other filler metals, is incapable of altering the chemistry of the heat-affected zone and zone of partial fusion. Thus attemps to combat the tendency to form carbide in the partial fusion zone and to form martensite in both the heat-affected zone and partial fusion zone have centered around control of the weld thermal cycle. One approach has been to recognize that both carbide and martensite result from rapid cooling. Therefore, very slow cooling can be beneficial. Among others, Rilley and Dodd[3] used high welding heat input and very large weld passes to retard cooling, thereby avoiding martensite and carbide formation. However, this becomes ineffective on large weldments when the mass of the structure acts as a quench. Alternately, preheats of 600° F were used by Bates[4] and others to prevent cracking, and a full post-weld anneal could then be used to remove carbide and martensite from the joint. However, annealing to eliminate carbides requires high temperatures and long periods of time which frequently result in considerable distortion.

More recently, Osman, et al.,[2, 5] Kotecki, et al.[6] and others have considered using the very low heat input of the short-circuiting transfer technique of gas metal-arc welding with nickel-based filler metal to limit carbides to discrete particles and martensite to a very thin layer along the weld fusion

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Table 1—Welding Conditions and Bend Test Results

Specimen	A	B	C	D	E	F	G	H	I	J	K	L	M	N	O	P	Q	R	S[a]	T[a]
Filler metal	0.062	0.062	0.062	0.062	0.045	0.045	0.045	0.045	0.045	0.045	0.062	0.062	0.062	0.062	0.062	0.035	0.035	0.035	0.035	0.035
diameter (in.) and designation	NS-61	NS-61	NS-61	NS-61	Imperial 75	Imperial 75	Imperial 75	Imperial 75	Imperial 75	Imperial 75	NS-61	NS-61	NS-61	NS-61	NS-61	NI-61	NI-61	NI-61	NI-61 0.062 NS-61	NI-61 0.062 NS-61
Average welding current, amp	280–310	280–310	280–310	280–310	185	185	185	185	185	185	280	280	280	280	280	160	160	160	160 300	160 300
Average welding potential, v	16–18	16–18	16–18	16–18	21	21	21	21	21	21	17	17	17	17	17	16	16	16	16	16
Electrode holder travel speed, ipm	50	50	60	60	50	60	70	80	80	80	60	83	109	100	100	83	83	83	83 40	83 40
Average heat input[b] (joules/in.)	6000	6000	5000	5000	4650	3890	3300	2910	2910	2910	4760	3440	2620	2860	2860	1850	1850	1850	1850	1850
Shielding gas and flow rate, cfh	Argon; 45	Argon; 45	Argon; 45	Argon; 45	CO_2; 30	CO_2; 30	CO_2; 30	CO_2; 30	CO_2; 30	CO_2; 30	Argon; 50	Argon; 50	Argon; 50	Argon; 50	Argon; 50	Argon; 30	Argon; 30	Argon; 30	Argon; 30	Argon; 30
Number of passes	26	26	34	34	1	1	1	1	1	1	1	1	1	1	1	1	1	1	28 & 17	28 & 17
Interpass temperature, °F	100	425–463	525–550	575–600	575–600	575–600	575–600	575–600	70	700	600	600	600	70	70	70	600	70	100	100
Postweld heat treatment, °F; hr	900; 2	None	None	None	None	None	None	None	1150; 1	None	None	None	None	1150; 1	None	None	None	1150; 1	None	1150; 1
Bend test type[c]	T	T	T	T	L	L	L	L	L	L	L	L	L	L	L	L	L	L	T	T
Average bend angle at fracture, deg	31	42	27	33	0	0	0	0	0	0	34	42	53	51	45	75	57	180	61	65

[a] Top number describes passes contacting ductile iron; bottom number describes passes used to fill in joint. If only one number is given, it applies to all passes.
[b] Passes contacting ductile iron only.
[c] "L" indicates longitudinal face bend; "T" indicates transverse side bend. All bend test specimens are ⅜ in. thick and were bent around a 1½ in. diam pin.

line. Then after a short postweld heat treatment (e.g., 1100° F for 2 hr) to temper the martensite and effect a stress relieved material, crack-free, machineable joints of considerable ductility were obtained. The feasibility of this technique, even without the postweld heat treatment, for both repair and fabrication of ductile iron was also reported by Fröberg.[7]

The Present Study

The present study is motivated by the consideration that, although the short-circuiting transfer technique has led to real progress in welding ductile iron, joint ductility and machineability have not usually been equal to that of ferritic ductile iron. Furthermore, the postweld heat treatments used to temper martensite may be objectionable in some applications.

Weldment preheat temperatures near or above the M_s temperature of the base metal have been used for preventing martensite formation in many ferrous materials during welding. This preheat slows the cooling rate so that the equilibrium transformation of austenite to ferrite and pearlite, or to bainite, occurs instead of martensite formation. Either bainite or the ferrite-pearlite mixture is more ductile than martensite.

One might therefore expect that preheat, which eliminates martensite from the weldment, would have a beneficial effect on the ductility of welds made via the short-circuiting transfer technique in ductile iron. However, an additional factor must also be considered. The success of the short-circuiting transfer technique depends upon its ability to limit carbides to discrete particles. The extent to which carbides form depends upon the length of time that the partial fusion zone remains in or above the eutectic temperature range of the ductile iron. The reason for this is that time determines the amount of carbon diffusion which produces the low melting point composition in the structure. Preheat would prolong this time, thereby aggravating the carbide problem while tending to eliminate the martensite problem.

Accordingly, the questions to be explored in this study are:

1. What minimum preheat temperature would eliminate martensite from a short-circuiting transfer type weldment in ferritic ductile iron?

2. To what extent will this preheat adversely affect carbide geometry in the partial fusion zone?

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Procedure

The equipment used included a 500 amp 100% duty-cycle constant potential power source, with variable slope and inductance, a side-beam carriage for automatic control of electrode holder travel speed, a hot plate capable of maintaining temperatures up to 950° F, and a fixture for maintaining joint alignment and restraint. Temperature ranges were determined by use of temperature sensitive crayons. The welding conditions used in making the various weldments are listed in Table 1. The conditions used for specimen A were arrived at during a previous study,[6] and those used for succeeding specimens[5] were determined as necessary to improve bend test results.

Three types of joint preparations (Fig. 1) were employed. Each single-pass weldment was prepared with the single U. Specimens A through D were prepared with the double vee, while specimens S and T were prepared with the double U. To afford better root accessibility later tests were made with a double U rather than a double vee. Each multipass weldment was back-gouged with a shaper before the second side was welded to insure complete root penetration.

The single-pass welds in shallow single U joint preparations were made to allow study of the fusion line microstructure without the interfering effects caused by the reheating due to multi-pass welding, and at the same time to permit bend testing of the specimen. The multi-pass welds were subjected to transverse side bend tests, and the results in Table 1 are averages of from two to four test specimens. The single-pass welds were subjected to longitudinal face bend tests (one per weld).

By way of comparison between the two types of bend tests, it might be noted that a longitudinal face bend requires simultaneous yielding of the weld metal, unaffected base metal, and the regions along the fusion line. Consequently, this bend test is the more severe for initiating cracks along the fusion line. On the other hand, while the transverse side bend test does not require immediate yielding along the fusion line as soon as the test is begun, it does orient the fusion line in such a way that cracks, once initiated, propagate readily along the fusion line. Bend angles were measured at the point of fracture except for those in the specimens made with mild steel. In these specimens, the weld metal

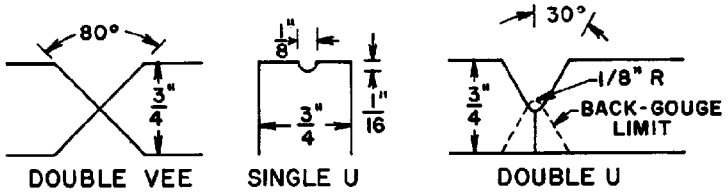


Fig. 1—Joint preparations used

was extensively macroscopically cracked before testing, and therefore the weldment can be said to have failed before bending.

The base metal used in this study was 3/4 in. thick ferritized ductile iron plate (grade 60-45-15) having a nominal carbon equivalent of 4.45.

Three welding wires were used in making the test weldments: International Nickel Company "NS-61," (97% nickel), INCO "Ni-61" (93% nickel), and Mid-States Steel & Wire Company "Imperial 75" (mild steel wire intended for gas metal-arc welding of mild steel).

In computing the heat inputs given in Table 1, the following formula was used:

$$\text{Heat input (joules/in.)} = \frac{60 \times \text{average arc voltage} \times \text{average arc amperage}}{\text{Electrode holder travel speed (ipm)}}$$

Results

Table 1 lists the welding conditions used in preparing each specimen and the bend angles obtained at fracture. By way of comparison, an unwelded sample of the base metal used in this study, having the same thickness as the weld specimens studied, bent 180 deg without fracture, although numerous small cracks could be observed on the surface of the sample following bending. This was used to represent the ultimate results of the welded material.

Specimens A through D provide partial answers to the questions posed earlier. Specimen A (Fig. 2) was given a postweld heat treatment at 900° F for one hour to temper martensite. The carbides in specimen A form around the graphite spheroids in the zone of partial fusion. They appear at or near the weld metal interface in Fig. 2 and are clearly discontinuous.

Specimen B was preheated into the range 425 to 463° F. Figure 3, taken in the zone of partial fusion, shows that this preheat was insufficient to prevent martensite formation. The martensite is clearly visible in the heat-affected zone. In addition, an increase in the amount of carbide over that of specimen A was apparent.

This increase in the amount of carbide prompted a reduction in welding heat input for specimens C and D. Specimen C also exhibited martensite indicating that a preheat of 525 to 550° F is insufficient to prevent martensite formation. This martensite was not eliminated until a preheat of approximately 600° F was employed for weldment D (Fig. 4A). Even in this sample, however, a few areas remain which contain some martensite.

Comparison of specimens A and D (Figs. 2 and 4B) emphasizes the adverse effect of preheat on carbide geometry. Even though specimen D was made with a lower welding heat input than specimen A, the carbides of specimen D are larger

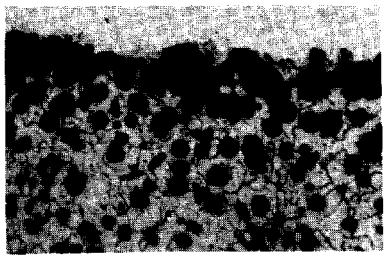
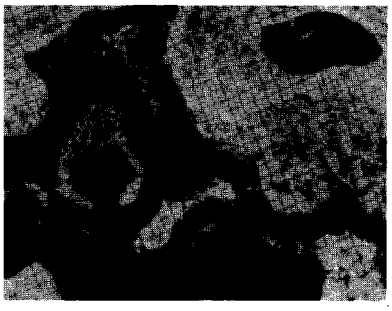


Fig. 2—Fusion line microstructure of specimen A made with an interpass temperature of 100° F, heat input of 6000 joules/in., and a postweld heat treatment at 900° F for 1 hr—nickel filler metal, 2% nital etch. X 50 (reduced 50% on reproduction)



Fig. 3—Fusion line microstructure of specimen B made with an interpass temperature in the range 425–463° F, heat input of 6000 joules/in., and no postweld heat treatment, nickel filler metal, 2% nital etch. X 200 (reduced 50% on reproduction)

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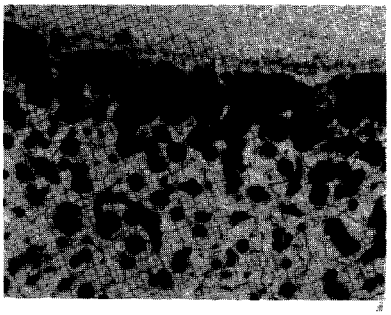


Fig. 4—Fusion line microstructure of specimen D made with an interpass temperature in the range 575–600° F, heat input 5000 joules/in., and no postweld heat treatment, nickel filler metal, 2% nital etch. A (top)—X 200; B (bottom)—X 50 (reduced 50% on reproduction)

and more nearly continuous than those of specimen A, and extend further from the weld proper. The bend angles obtained at fracture however, are similar—Table 1.

Since it has been indicated that a preheat of 600° F is sufficient to prevent martensite formation in the heat-affected zone, a logical question at this point is whether or not this same preheat will prevent martensite formation in the weld proper if the filler metal were mild steel instead of nickel. Accordingly, specimens E through H were made with a constant preheat of 575–600° F and with decreasing heat input. Each exhibited extensive cracking in the weld metal immediately after welding, and microscopic examination revealed that martensite was present throughout the weld proper. Likewise, specimen I, made without preheat, exhibited cracking and martensite throughout the weld proper.

For specimen J, using a mild steel filler metal, the preheat temperature was increased to 700° F, with a welding heat input of 2910 joules/in. From the microstructure along the fusion line (Fig. 5) two observations are made. First, comparison of the fusion line carbides with those of specimen D (Fig. 4B) shows a considerable increase in the extent of carbide



Fig. 5—Fusion line microstructure of specimen J made with a preheat of 700° F, heat input of 2910 joules/in., mild steel filler metal, 2% nital etch. X 50 (reduced 50% on reproduction)

formation at the higher preheat temperature, despite the reduction in welding heat input from 5000 to 2910 joules/in. for the latter weldment. Second, the weld proper exhibits a dual structure—Fig. 5. The regions containing coarse martensite exhibit a great tendency to initiate and propagate cracks.

The formation of martensitic regions in the weld proper despite a preheat of 700° F indicates a higher hardenability of the weld proper. Since the weld proper consists of a mixture of high carbon iron from the base metal and mild steel from the filler metal, this higher hardenability appears attributable to the presence of deoxidizers from the mild steel welding wires. The nominal composition of this filler metal includes 1.08% manganese and 0.55% silicon.

The increase in the extent of carbide formation observed in specimen J indicated that further exploration of still higher preheat temperatures would be of little value.

The observed adverse effect of the 600° F preheat on carbide geometry in specimens A through D prompted study of welding heat inputs of less than 5000 joules/in. to combat this effect. Specimens K, L, and M were made with decreasing heat input (from 4760 down to 2620 joules/in.), each with 600° F preheat. Improved carbide geometry with decreasing heat input is reflected by increasing bend angles at fracture—Table 1.

Specimen M was welded at 109 ipm. This welding speed appeared to be the maximum at which nickel would consistently wet ductile iron under the conditions listed for specimen M in Table 1. Without preheat, the maximum electrode holder travel speed under otherwise similar conditions appears to be 100 ipm, giving a heat

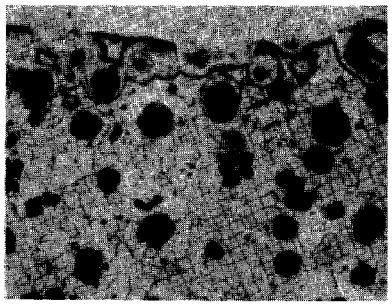


Fig. 6—Fusion line microstructure of specimen P made with no preheat nor postheat, 1850 joules/in. heat input, nickel filler metal, 2% nital etch. X 100 (reduced 50% on reproduction)

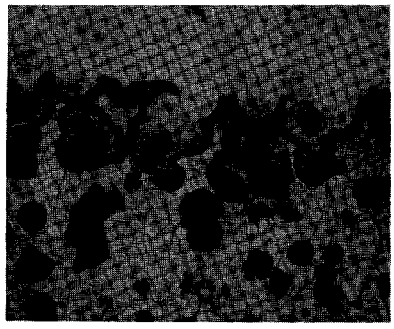


Fig. 7—Fusion line microstructure of specimen Q made with 600° F preheat, no postheat, 1850 joules/in. heat input, nickel filler metal, 2% nital etch. X 100 (reduced 50% on reproduction)

input of 2860 joules/in. Specimens N and O were made under these conditions, and N was postweld heat treated while O was not. The beneficial effect of postweld heat treatment is reflected by the bend angles at fracture—Table 1.

In order to reduce the welding heat input still further, smaller diameter nickel filler metal was used (0.035 vs. 0.062 in.). Specimens P, Q, and R were produced with this finer filler metal, under the conditions listed in Table 1. In particular, a heat input of 1850 joules/in. was used. Efforts to reduce the heat input still further by increasing electrode holder travel speed encountered difficulty in obtaining wetting of the nickel on the ductile iron without preheat. Accordingly, the arc conditions were held constant for these three specimens in order to permit comparison.

The fusion line microstructures of specimens P (no preheat nor postheat) and Q (600° F preheat and no postheat) are compared in Figure 6 and 7. In specimen P (Fig. 6) the carbides are confined to very thin "halos" encircling their parent graphite spheroids, but may

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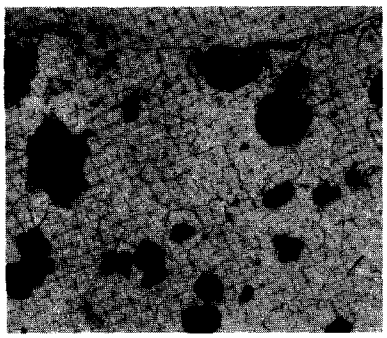


Fig. 8—Fusion line microstructure of specimen R made with no preheat, 1150° F postheat, 1850 joules/in. heat input, nickel filler metal, 2% nital etch. X 100 (reduced 50% on reproduction)

be distinguished from the untempered martensite which in turn surrounds them. The carbides of specimen Q (Fig. 7) are considerably larger than those of specimen P, though still discontinuous, and extend further from the weld proper. Furthermore, the partial fusion zone plus the heat-affected zone, in the weldment made without preheat, weldment P (Fig. 6) is no thicker than the partial fusion zone in the weldment made with preheat (weldment Q of Fig. 7). This is reflected in a comparison of the bend angles at fracture for the two specimens—Table 1. The weldment made without preheat and without postheat could be bent considerably further than that made with preheat (75 vs. 57 deg).

Specimen R, made without preheat and given a postweld heat treatment, could be bent 180 deg without fracture, apparently equalling the unwelded base metal, although the cracks originating from the fusion line of the weldment are larger than those in the base metal after bending It should be noted, however, that the comparison of this behavior in bending with that of specimens P and Q may be misleading, because the base metal of specimen R came from the edge of one of the cast plates, where the nodule count is considerably lower than that of specimens P and Q—Fig. 8. As a result of the low nodule count, carbides in specimen R are much more widely separated than in the other two specimens.

Since multipass weldments are of more engineering interest than single-pass weldments, multipass butt welds were then made using the fine wire to join the halves of the weldments at the root and butter the sides of the joint preparation. The coarser nickel filler

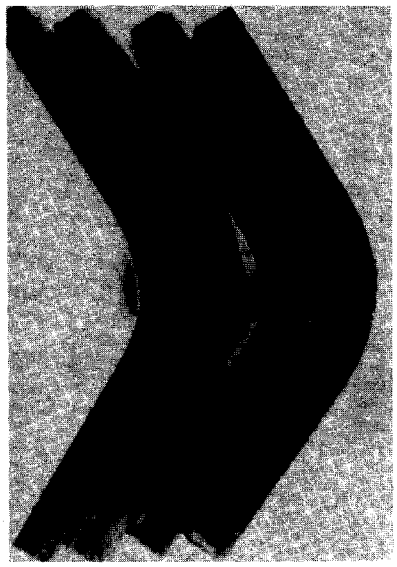


Fig. 9—Transverse side bend test specimens of weldments S (left) and T (right). X 3/4 (reduced 13% on reproduction)

metal was used to complete the joints. Resulting specimens S and T are also described in Table 1. Both were made without preheat, while specimen T was given a postweld heat treatment at 1150° F.

Transverse side bend test results for these two specimens are similar, with the postheated specimen exhibiting somewhat greater bend angles (65 vs. 61 deg)—Fig. 9. This similarity is probably, in part, due to the reheating effect of multipass welding—the martensite formed by one welding pass being tempered to a certain extent by the succeeding passes.

The fusion line microstructures of these two weldments are similar. In both weldments, the fusion line carbides are small and very discontinuous, except for a few regions in which it appears that the partial fusion zone was remelted by the passes with the coarse wire (Fig. 10). These remelted areas are present in both specimens. Therefore, it appears that comparison of their bend angles is valid, although neither represents the optimum attainable.

Summary

The stated aim of this study was to explore two questions:

1. At what preheat temperature will martensite formation be prevented when using the short-circuiting transfer technique when welding ASTM grade 60-45-15 ductile iron?

2. How does preheat affect carbide geometry?

It has been indicated that a pre-

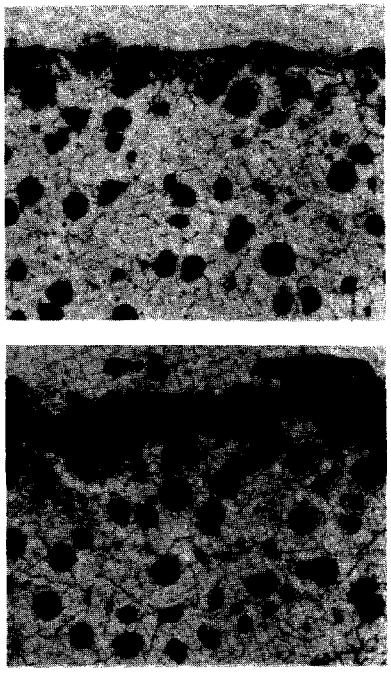


Fig. 10—Fusion line microstructure of specimen T. A (top)—where remelting did not occur; B (bottom)—where remelting did occur. 2% nital etch. X 100 (reduced 50% on reproduction)

heat temperature of approximately 600° F is sufficient to prevent martensite formation in the heat-affected zone of these welds. It has also been shown that a preheat of 700° F is not sufficient to prevent martensite formation in the weld proper when a given commercial mild steel filler metal was used.

Second, it has been shown that a 600° F preheat considerably increases carbide size and continuity over that obtained when the base metal is at or near room temperature before welding. This adverse effect can be partially compensated for by reducing the welding heat input within the limits imposed by the nickel wetting ductile iron. At very low welding heat inputs, on the order of 2000 joules/in. or less, this effect appears sufficiently adverse as to make welding with no preheat nor postheat seem preferable to using preheat to prevent martensite formation.

Whether or not preheat was used, in no case was heat-affected zone cracking observed in connection with nickel filler metal. The extreme thinness of the heat-affected zone apparently prevents stress levels from becoming high enough to cause cracking.

It should not be assumed, however, that a 600° F preheat will prevent martensite formation in pearlitic grades of ductile iron.

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Hardenability of ferrous material depends strongly upon local carbon content. In pearlitic grades of ductile iron, as soon as austenite is formed in the heat-affected zone, it will have a relatively high carbon content due to the dissolution of the fine carbide lamellae of the parent pearlite—the diffusion distance between lamellae is extremely short. On the other hand, in ferritic ductile iron, the carbon which can increase hardenability must first diffuse outward from the graphite spheroids. The distance between these spheroids is several orders of magnitude greater than the distance between pearlite lamellae. Consequently, for the very short times at elevated temperature associated with the short-circuiting transfer technique of welding, pearlitic ductile iron is likely to have greater hardenability than ferritic ductile iron. Also, alloying elements added to ductile iron are likely to increase the local hardenability.

Finally, when very low heat input is used for the short-circuiting transfer technique of buttering or surfacing, the resulting deposit has a small cross-sectional area. Thus in multipass welding, some limitation of the welding heat input is also required for passes adjacent to the buttering (or surfacing) to prevent remelting of the base metal partial fusion zone.

Acknowledgments

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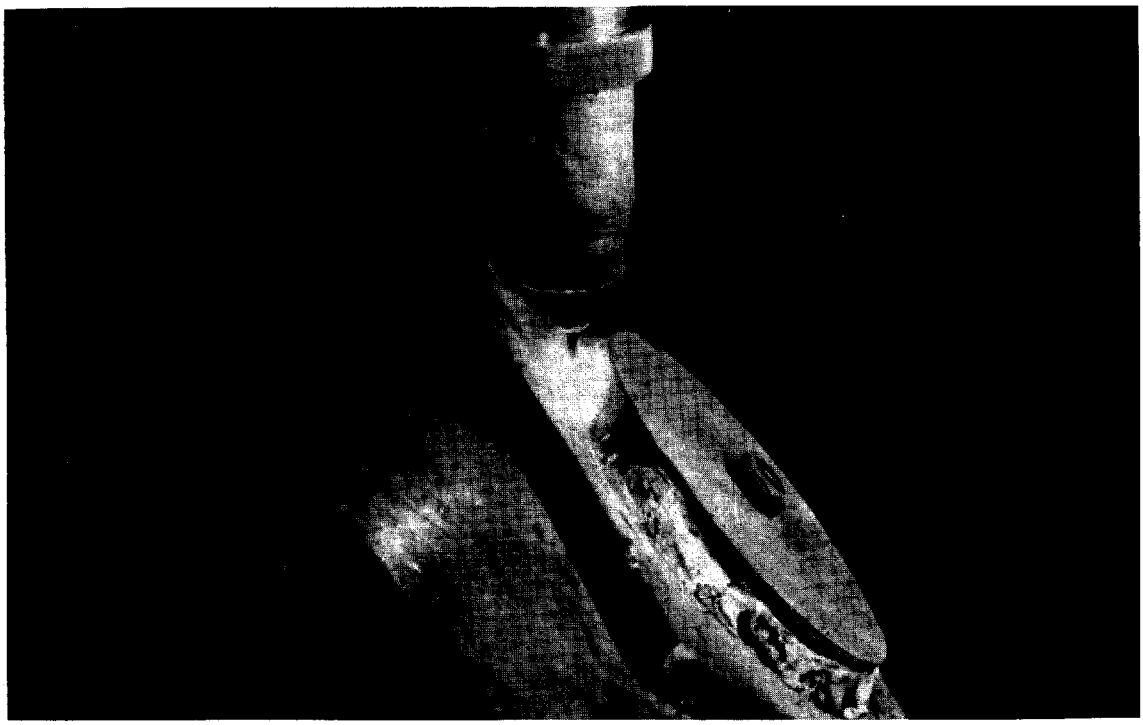
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A ductile iron disc being welded to a carbon steel plate by automatic flux-cored arc welding

Flux-Cored Electrode for Cast Iron Welding

BY R. A. BISHEL

presents the first new development in many years for the welding of cast iron

ABSTRACT. Ductile and gray irons are being increasingly used for cast parts. In a variety of applications in assembled equipment such as farm machinery, construction equipment, and machine tools these cast irons provide good service at low cost. In many cases, however, the extent to which the cast irons have been used has been limited by the lack of a high-deposition-rate, automatic welding method. The development of a new filler metal in which the concept of the flux-cored electrode is applied to cast-iron welding, promises to remove that limitation. The new flux-cored electrode uses a nickel-iron sheathing material with carbon, slagging ingredients, and deoxidizers added to the core. The product has produced high-efficiency joints in ductile iron and gray iron and has been successfully used to join ductile iron to several dissimilar materials. This paper describes the results of a testing program conducted to define operating parameters for this new electrode and to evaluate various types of joints welded with it.

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Introduction

The need for an automatic welding method for cast irons has existed for many years. The availability of such a method would greatly reduce the cost of cast-iron structures assembled by welding. It would also enable many forged and cast-steel parts to be replaced by parts of lower cost gray or ductile iron, and it would permit some complex castings to be redesigned as low-cost welded assemblies. In some instances, parts being produced as one-piece forgings could be produced in sections, with forgings used only where necessary and remaining sections replaced by castings.

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Previously available methods for joining, repairing, and surfacing ductile and gray cast irons include brazing, braze welding, oxyacetylene welding with cast-iron rod, and shielded metal-arc welding with flux covered electrodes such as ENi-CI or ENiFe-CI. These processes have low deposition rates, and they are unsuitable for automatic or semi-automatic operations. Short-circuiting gas metal-arc welding with filler metals such as ERNi-3 and ERNiCu-7 has also been used, but with little success. Of these processes, shielded metal-arc welding with ENi-CI or ENiFe-CI electrodes produced the highest quality joints.

The new flux-cored electrode, designated NI-ROD* FC 55, is the first welding product specifically designed for high-speed production welding of cast irons. Weld metal deposited by this electrode has a chemical composition similar to that deposited by the ENiFe-CI covered electrode, that is, about 50% nickel, 49% iron, and 1% carbon. The cored electrode can be used without shielding gas, with CO_2 or CO_2 plus argon shielding, or with submerged arc flux. Standard equipment for flux-cored arc welding (usually the same equipment used for gas metal-arc welding) or submerged arc welding is used.

Operating Characteristics

Flux-cored arc welding of cast irons is a new area in the field of welding. Much of the development work on FC 55 involved defining the product's operating characteristics and establishing welding parameters for various joint types. The objective was to accumulate the basic information needed to apply the process to specific commercial operations.

Effects of Casting Composition and Defects

As with any welding process, the first item for consideration is the condition of the base metal. Potentially detrimental elements such as sulfur, phosphorus, and lead are normally held to acceptable levels during production of castings. However, abnormally high levels of such elements may cause welding difficulty or defective welds. High levels of phosphorus are particularly detrimental. Depending on the amount of weld-metal dilution, phosphorus may cause hot-cracking of the weld.

The casting must be sound and free of inclusions and spongy areas. Such defects can cause severe porosity in the weld. Spongy areas may also be high in phosphorus or some other harmful element. The best method to detect casting defects is by x-ray inspection.

Castings may also be impregnated with oil. If not removed, the oil will cause porosity in the weld metal. Oil can be removed from castings by a heat treatment of 900-1000F (480-540C) for a few minutes.

The as-cast surface, or skin, should be removed by grinding or machining before welding is begun. Unremoved casting skin will cause slag sticking, and it may adversely affect weld quality.

Preheat

Preheat is not usually necessary for ductile iron, but better wetting action (wider weld bead) can be obtained with the use of preheat.

With gray iron, preheat is helpful in preventing fusion-zone cracking. The preheat temperature should be at least 400F (205C) and preferably over 600F (315C).

Shielding

The core of FC 55 contains deoxidizers, and the electrode is designed to be operated with an open arc (no shielding gas). A shielding gas may be desirable for moisture exclusion under conditions of high humidity or for greater protection when a longer arc (higher voltage) is used to increase wetting by the molten weld metal.

Carbon dioxide, argon, and nitrogen were evaluated as shielding gases. The best results were obtained with CO_2. Some mixtures of argon and CO_2 also gave good results. Argon, however, showed no apparent advantage and at higher concentrations (over 30%) caused slag sticking. Similar results were obtained with nitrogen. It offered no improvement over CO_2 and caused slag sticking.

The cored wire can also be operated with submerged arc flux. Of the fluxes tested, a proprietary mixture (INCOFLUX 5 Submerged Arc Flux) produced the best results. Weld metal properties were equivalent to or higher than those obtained with CO_2 shielding or an open arc, and bead contour and deposition rate were similar. The main advantage of submerged arc flux is that no smoke is produced by the arc. The absence of smoke could be important from the standpoint of OSHA if smoke-extracting equipment is not available.

Power Sources

Direct current, pulsing arc, and alternating current power sources were evaluated for use with FC 55. Direct current power sources with either constant voltage or constant current and with reverse polarity gave the best results. Most of the work reported here was done with reverse polarity direct current using a constant voltage power source.

A pulsing arc power source was also used successfully. With the pulsing arc, lower current and voltage levels could be obtained, and arc blow was decreased.

Current

A wide range of current was used with the flux-cored electrode. Current varied from 200 A with the pulsing arc power source to 450 A with the direct current, reverse polarity source. The optimum current range for moderate travel speeds was found to be 330 to 380 A. Currents up to 450A were required for travel speeds over 35 in./min (89 cm/min). At a given travel speed, higher currents produced deeper penetration. Figure 1 shows the effects of current on weld beads deposited with a direct current, reverse polarity power supply. Figure 2 shows beads deposited at various currents with a pulsing arc power source.

Current also influenced deposition rate. At 250 A, deposition rate was 9 lb/hr (4 kg/hr); at 380 A, deposition rate increased to 15 lb/hr (6.8 kg/hr).

Voltage

Voltage is related to arc length (distance between the workpiece and electrode). Arc length increases as volt-

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Table 1 — Effect of Voltage on Weld Beads Deposited[a] on Ductile Iron

Voltage, V	Bead width, in.	Bead width, cm	Penetration, in.	Penetration, cm	Weld area[b], in.²	Weld area[b], cm²	Dilution, %	Carbon %[c]
25	0.52	1.32	0.080	0.203	0.068	0.433	44	1.84
28[f]	0.54	1.37	0.065	0.165	0.080	0.516	38	1.62
28	0.50	1.27	0.060	0.152	0.063	0.406	46	2.01
30	0.56	1.42	0.080	0.203	0.076	0.490	49	1.94
30[d]	0.58	1.47	0.100	0.254	0.078	0.503	55	2.13
30[e]	0.62	1.57	0.085	0.216	0.085	0.548	56	2.11
32	0.58	1.47	0.070	0.178	0.070	0.451	54	2.05
34	0.60	1.52	0.090	0.229	0.073	0.471	58	2.15
35	0.60	1.52	0.060	0.152	0.066	0.426	56	2.09

(a) Welding conditions: current, 350 A; travel speed, 20 in./min (51 cm/min); electrode extension, 1 in. (2.54 cm); gas, 70 cfh (2 m³/hr) CO ; polarity, reverse.
(b) Bead cross-section.
(c) Carbon content of weld bead calculated from dilution rate and carbon contents of base metal and electrode.
(d) Base metal preheated to 400 F (205 C).
(e) Base metal preheated to 600 F (315 C).
(f) Straight polarity.

age increases. Greater arc lengths increase the tendency for air to mix with the molten weld metal passing from the electrode to the workpiece. The air can consume the deoxidizers in the core, causing porosity in the weld metal. With open-arc welding, voltages above 31V can cause porosity. Shielding gas can be used to prevent porosity at the higher voltages.

As shown in Fig. 3, bead width also increases with voltage. The arc column has the shape of a cone, with the vertex at the electrode. The base of the cone increases in area as voltage increases, resulting in a wider bead.

The effects of voltage variations on bead characteristics are summarized in Table 1.

Travel Speed

The flux-cored electrode was operated at travel speeds ranging from 10 in./min (25 cm/min) to 45 in./min (114 cm/min). At 350 A and 30V, the optimum travel speed was 20 in./min (51 cm/min). Higher travel speeds usually required higher current, especially for fillet welds.

Electrode Extension

The amount of electrode extension (in this study, the distance from the contact tip to the base plate or previous bead) affected the degree of penetration by the arc. With 0.093-in. diam (0.24 cm) electrode, 350 A, and 30V, the optimum electrode extension was 0.75 in. (1.9 cm). Shorter electrode extensions caused deeper penetration. Extensions less than about 0.5 in. (1.3 cm) increased the tendency for slag entrapment. Longer electrode extensions decreased penetration. Figure 4 illustrates the effect of electrode extension on penetration.

As electrode extension is increased the wire be-



Fig. 1 — Effect of current on bead shape and penetration. All beads were deposited at the same travel speed and voltage. Amperage values were, from left to right, 200, 200, 250, 300, 350, 400, and 500. Power supply was direct current, reverse polarity



Fig. 2 — Effect of current on beads deposited with a pulsing-arc power supply. All beads were deposited at the same travel speed with peak voltage at 45 V and background voltage at 37V. Amperage values were, from left to right, 250, 240, 230, 275, 285, 295, 310, 330, 340, and 350

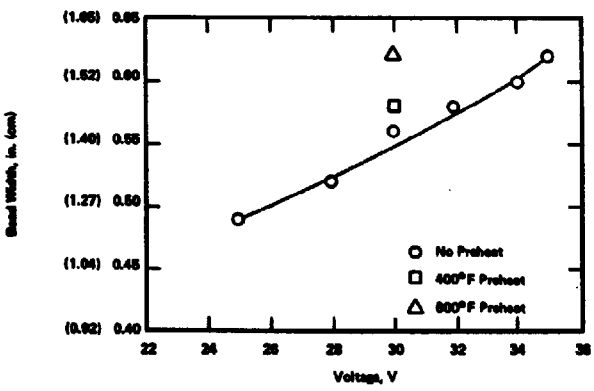


Fig. 3 — Effect of voltage on width of beads deposited on ductile iron

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comes hotter (the I^2R relationship). A given amount of power is available at a given current and voltage. If the power is used to heat the wire, less power is available to penetrate the workpiece. If the electrode is overheated, its sheath will melt. The molten ball will twirl around the end of the electrode and be thrown from the welding area by the magnetic field, causing spatter.

Deposition Rate

The deposition rate obtained with FC 55 varied with welding current. As shown in Fig. 5, deposition rates of over 15 lb/hr (6.8 kg/hr) were achieved at the higher current settings.

The deposition efficiency of FC 55 was found to be about 80%. That is, for each pound of electrode, 0.8 lb will deposit as weld metal. The 20% loss represents slagging ingredients, deoxidizers, and spatter.

Multiple Passes and Weave Beads

With the single pass technique, structures in the heat-affected zone were martensite, carbide, or a combination of both. Multipass welding was beneficial because subsequent passes tempered the heat-affected zones produced by previous passes. The structure in the heat-affected zone was then tempered martensite, or bainite. Carbides were not changed by subsequent passes.

Both weave beads and stringer beads were deposited with FC 55. Weave beads up to 1.5 in. (3.8 cm) wide were produced. The two techniques produced equivalent weld metal properties. Weaving, of course, decreased travel speed.

Out-of-Position Welding

The high deposition rates of FC 55 normally made out-of-position welding impractical. Horizontal welds, however, were successfully made with small diameter (0.078 in.) (0.19 cm) wire and a pulsing arc power source.

Slag Entrapment

As with any welding process that produces slag, slag entrapment can occur in flux-cored arc welding of cast

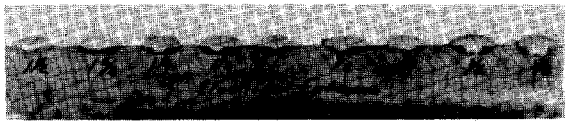


Fig. 4 — Effect of electrode extension on penetration. Numbers under beads indicate electrode extensions in inches. Beads were deposited on steel at 350A, 30V, and 20 in./min (51 cm/min) travel speed

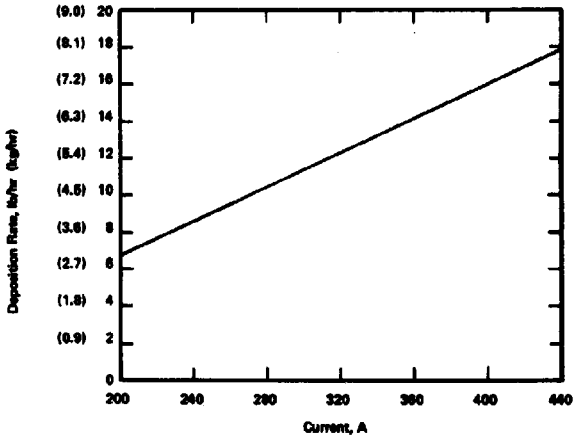


Fig. 5 — Effect of welding current on deposition rate of FC 55

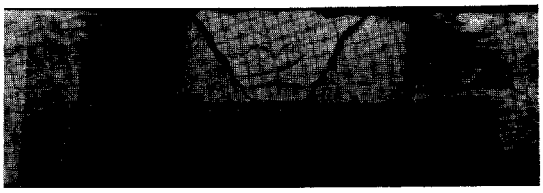


Fig. 6 — Slag entrapment under weld beads in ductile iron Vee-groove weld

Table 2 — Typical Tensile Properties of Groove Joints in Ductile Iron

Specimen	Shielding	Yield Strength (0.2% Offset) ksi	MPa	Tensile strength, ksi	MPa	Elongation, %	Reduct. area, %	Hardness, Rb
All-weld-metal	None	45.0	310	69.0	476	15.5	14.5	81
All-weld-metal	CO_2	45.5	314	72.0	496	21.0	18.8	80
All-weld-metal	Sub-arc flux	49.0	338	74.0	510	18.5	20.6	86
Transverse	None	43.5	300	66.0	455	—	—	—
Transverse	CO_2	44.0	303	66.0	455	—	—	—
Transverse	Sub-arc flux	45.0	310	64.0	441	—	—	—
All-weld-metal[a]	CO_2	44.0	303	68.0	468	15.0	16.2	80
Transverse[a]	CO_2	43.5	300	67.7	467	—	—	—

(a) Pulsing-arc power source.

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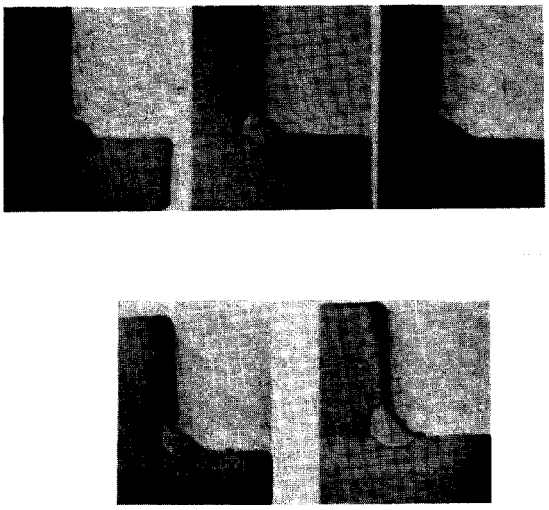
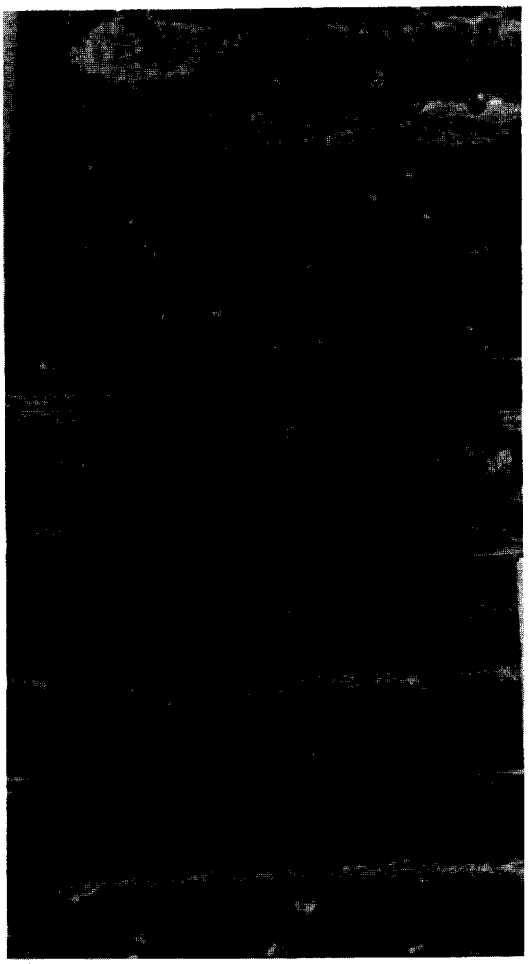


Fig. 7 — Undercutting in fillet welds between ductile iron and steel. Joints in top row were welded at 350 A and 30 V with travel speeds, from left to right, of 35, 40, and 45 in./min (89, 102, and 114 cm/min). Joint at lower left shows increased undercutting that resulted from an increase in voltage to 35 V at 350A and 35 in./min. Undercutting was eliminated (lower right) with settings of 450 A, 33 V, and 35 in./min



irons. Slag entrapment can be controlled or eliminated by proper selection of parameters and proper bead placement.

Figure 6 shows slag entrapment in four areas of a welded joint. Slag entrapment occurred under Bead 2 because of a wide weave and slow travel speed. Under Beads 5 and 8, slag entrapment occurred because of improper bead placement and insufficient travel speed. A travel speed of around 20 in./min (51 cm/min) was found to minimize the possibility of slag entrapment during multipass welding.

The experience and skill of the operator are important factors in avoiding slag entrapment. A trained operator will know where to position the electrode and which parameters to vary. Grinding is the best method to remove areas that contain entrapped slag.

Undercutting

Undercutting creates a notch in the heat-affected zone, where martensite and carbide are present, and can decrease the service life of the joint. Undercutting along a groove face can entrap slag. Undercutting can be avoided by proper welding conditions.

Figure 7 shows the effects on undercutting of variations in parameters used for fillet welds between ductile iron and steel. Travel speed was varied from 10 in./min (25 cm/min) to 45 in./min (114 cm/min) with current maintained at 350A and voltage at 30V. Undercutting was encountered at a travel speed of 25 in./min (64 cm/min), and the degree of undercutting increased with higher speeds. Attempts to decrease undercutting by raising voltage resulted instead in more severe undercutting. An increase in current to 450 A at 35 in./min (76 cm/min) and 35V decreased undercutting considerably but did not eliminate it. At the same travel speed and current, voltage was lowered from 35V to 33V. Those settings eliminated the undercutting.

In other work, it was found that preheating the cast irons to temperatures over 400 F (205 C) reduced the tendency toward undercutting.

Spatter

Spatter occurs when the force of magnetic fields created by direct current flowing through the electrode exceeds the forces needed to transfer molten metal across the arc. Spatter is undesirable because it reduces deposition efficiency and, if it must be removed, increases the time required for completion of the joint.

Spatter from the flux-cored electrode could be controlled by adjustments in welding parameters. The amount of spatter increased with voltage. The transition voltage gap (from minimal spatter to significant spatter) for FC 55 was found to be between 30 and 32V. Figure 8 illustrates the effect of voltage on spatter.

Excessive electrode extension also caused spatter. The optimum electrode extension for control of spatter was about 0.75 in. (1.9 cm).

Fumes and Smoke

As do other flux-cored electrodes, FC 55 emits fumes and smoke during operation. To prevent operator dis-

Fig. 8 — Weld-metal spatter produced at various voltage levels. Welding voltages were, from top to bottom, 35 V, 34 V, 30 V, 28 V, and 25 V

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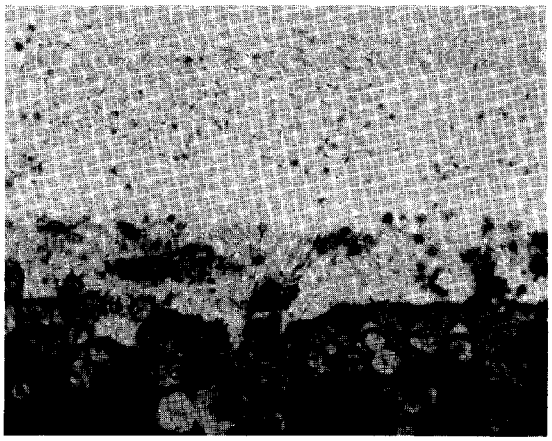


Fig. 9 — As-welded structure in heat-affected zone of joint in ductile iron. X100, reduced 30%

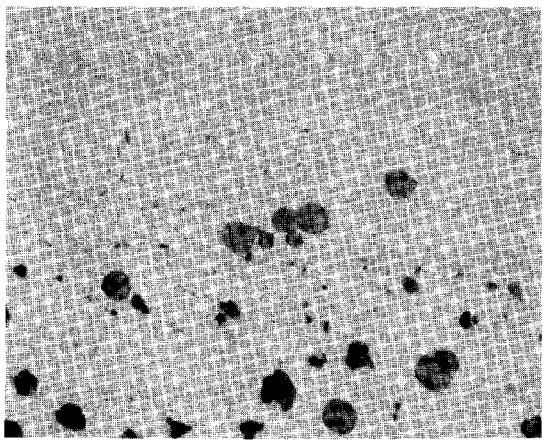


Fig. 10 — Heat-affected zone of joint in ductile iron after a ferritizing anneal. X100, reduced 30%

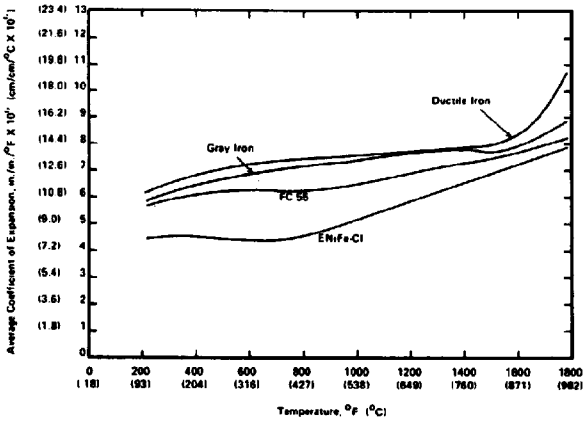


Fig. 11 — Average thermal-expansion coefficients (from 200 F (93 C) to temperatures shown) for cast irons and weld metals

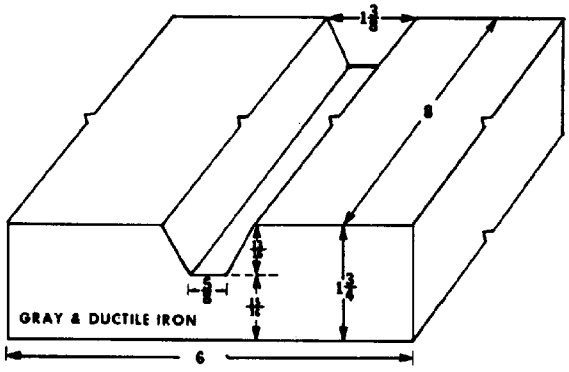


Fig. 12 — Test block used for evaluation of groove joints

comfort or obstruction of the operator's vision, smoke-extracting guns or ventilating systems should be used. Because of the self-shielding nature of the cored electrode, smoke-removal systems can be placed near the welding area.

Postheat

Postheating and slow cooling of joints in gray iron are beneficial in that they reduce the tendency of gray iron to crack in the fusion zone.

Postheating of ductile iron at temperatures of about 1000 F (540 C) greatly decreases the hard structures in the heat-affected zone.

Ferritizing Anneal

Although the extended times at high temperatures often make it impractical, the ferritizing anneal is one method to rid the heat-affected zone of troublesome carbides and martensite. The recommended treatment is 1550-1650 F (845-900 C) for 3 hr, furnace cool to 1275 F (690 C) and hold for 5 hr, furnace cool slowly to 1100 F (595 C), and then air cool to room temperature. As shown in Figs. 9 and 10, the ferritizing anneal decomposes the carbides and eliminates the martensite.

Coefficient of Expansion

As shown in Fig. 11, the coefficient of expansion for FC 55 weld metal is near the coefficients for gray and ductile iron. Similar expansion characteristics are necessary to prevent excessive stress levels during welding.

Evaluation of Welded Joints

Joints in ductile and gray iron were welded with FC 55 by various methods and tested for soundness and mechanical properties. Some joints between ductile iron and selected dissimilar metals were also evaluated.

Joints in Ductile Iron

Grade 60/45/10 ductile iron was used for the tests. Self-restraining groove joints were machined in 8 × 6 × 1.75 in. (20.3 × 15.2 × 3.4 cm) test blocks. The grooves were 0.75 in. (1.9 cm) deep. Figure 12 shows the configuration of the groove joints.

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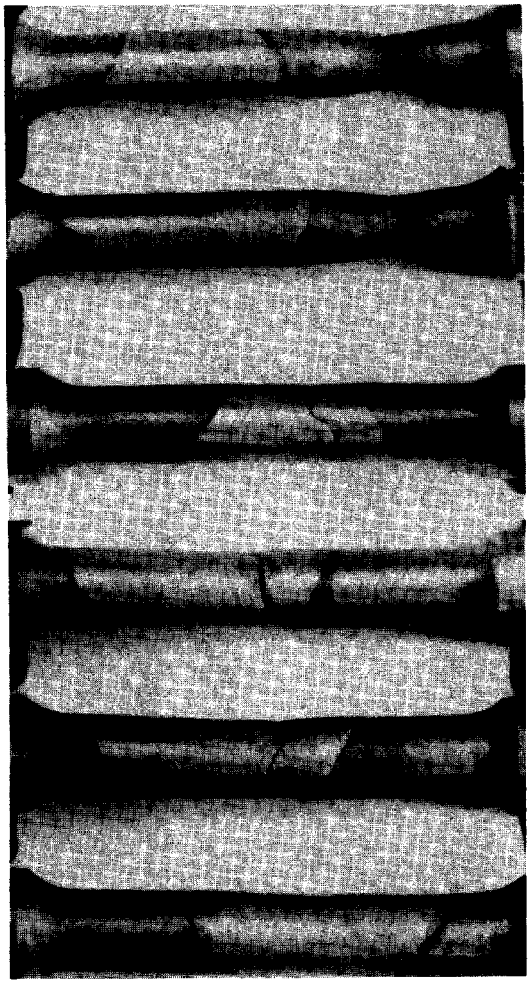


Fig. 13 — Failed transverse tensile specimens from joints in ductile iron

The groove joints were welded with an open arc, with CO_2 shielding gas, and with submerged arc flux. Typical welding conditions were:

Voltage	30V
Current	350 A, dc
Polarity	Reverse
Electrode Extension	0.75 in. (1.9 cm)
Travel Speed	20 in./min (51 cm/min)

Transverse and all-weld metal tensile specimens were cut from the completed test blocks. Slices for macro- and micro-examination were also prepared from the blocks. Some of the test blocks were ground and liquid penetrant inspected to check for weld metal cracking. Transverse slices were also liquid penetrant inspected and examined under a microscope.

The tensile-test results are given in Table 2. Figure 13 shows some typical failed transverse specimens. In most cases, failure occurred in or near the heat-affected zone. Specimens that failed in that area had tensile strengths up to about 70,000 psi (483 MPa). Some of the transverse specimens failed in the base metal, in which case the tensile strength was usually greater than 72,000 psi (496 MPa). From these results, it appears that the heat-affected zone is not the weakest area if it is narrow.

Preheating of the ductile iron to temperatures of 400F (205C) and 600F (315C), although it provided better wetting by the weld metal, did not improve soundness or strength of the joint. For some applications, preheating may be a disadvantage since it results in a wider heat-affected zone and increases the amount of carbides and martensite in the fusion zone.

Very little weld-metal cracking occurred in the groove joints. The test results indicate that if a wide root bead with extreme dilution is deposited, the resulting high carbon content of the weld metal could cause cracking. The carbon content required for cracking would depend on the stresses present, but it is estimated to be above 2.5%.

Some fillet joints in ductile iron were also evaluated. At 350 A and 30V, good results were obtained with

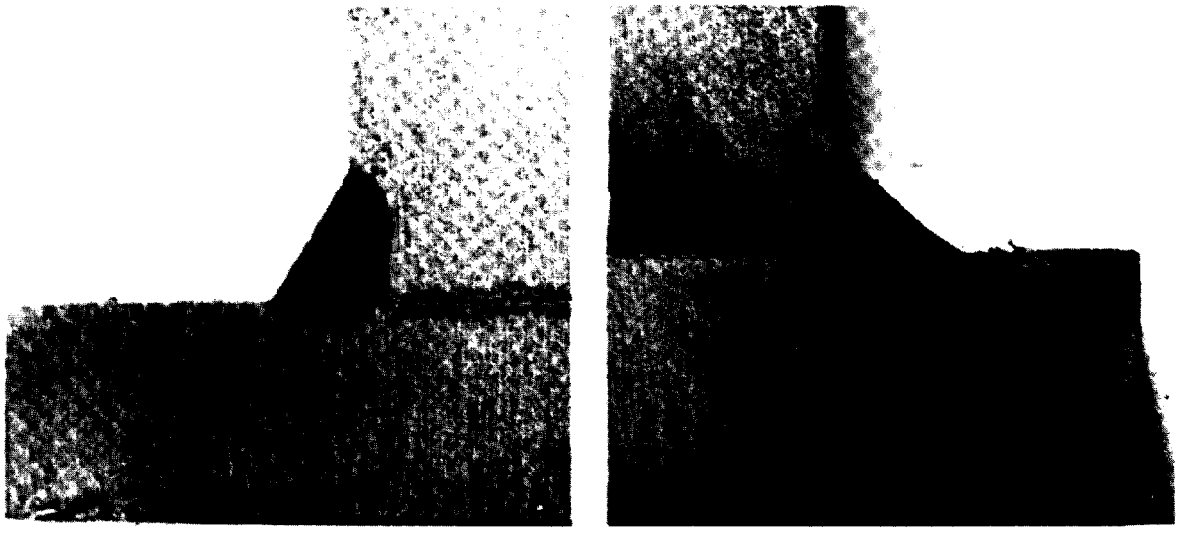


Fig. 14 — Fillet welds in ductile iron. Joint on left was welded at 30 in./min (76 cm/min) with 450 A and 32 V. Joint on right was welded at 15 in./min (38 cm/min) with 350 A and 30 V

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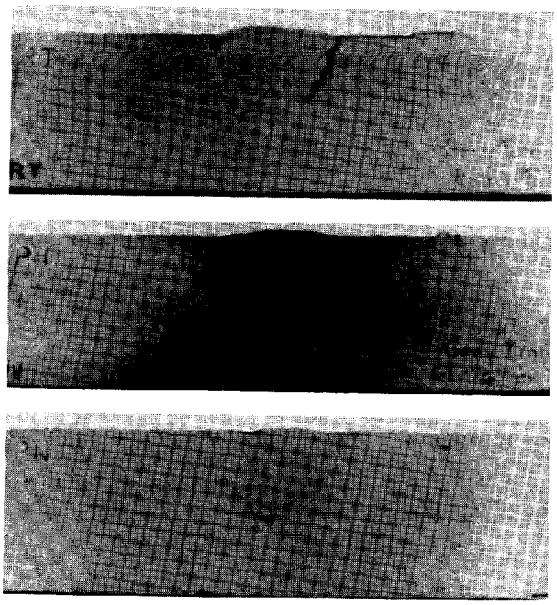


Fig. 15 — Effects of preheat and peening on fusion-zone cracking of gray iron. Top joint was completed without preheat or peening. Center joint was preheated to 600-650 F (315-345 C), but was not peened. Bottom joint was welded without preheat, but each bead was peened

travel speeds from 10 in./min (25 cm/min) to 25 in./min (64 cm/min). With travel speeds from 25 in./min to 45 in./min (114 cm/min), an increase in current to 450 A was necessary to prevent undercutting. Figure 14 shows cross-sections of fillet welds made at two different travel speeds.

Joints in Gray Iron

The low strength and low ductility of gray cast iron make the material particularly susceptible to cracking during welding. Shrinkage stresses that develop during cooling of the weld metal tend to separate the weld bead from the base metal. The hard structures (carbides and martensite) present in the heat-affected zone make that area the most vulnerable. High weld-metal strength is not a requirement for joints in gray iron.

The carbon added to the core of FC 55 forms dendritic and nodular graphite in the weld metal. The resulting increase in volume decreases shrinkage stresses. In addition, the coefficient of expansion of FC 55 weld metal is near that of gray iron. Even with these favorable circumstances, sufficient stress may be present to cause cracking in the fusion or heat-affected zone. The prevention of such cracking is the major factor governing welding procedures for gray cast iron.

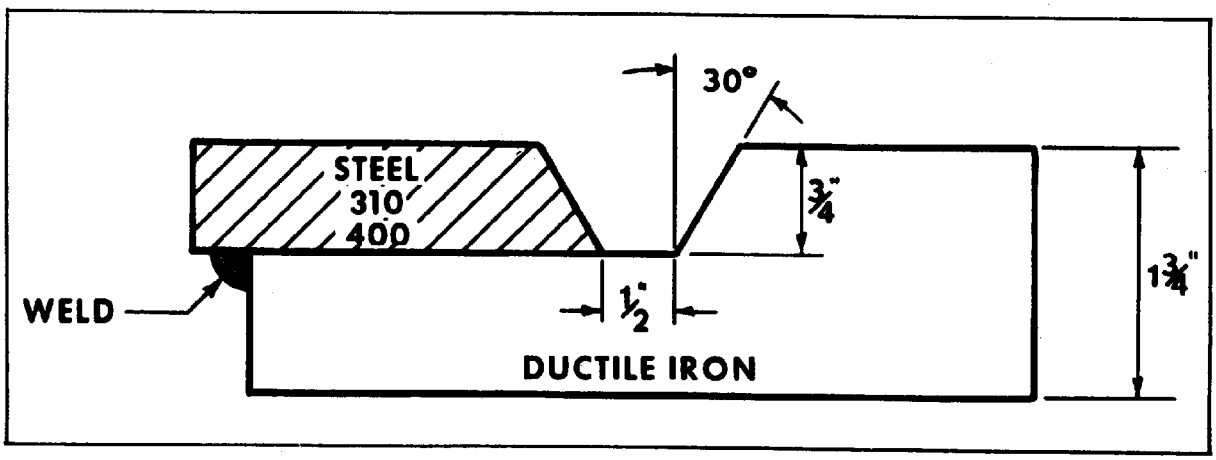
Gray iron also exhibits a greater tendency toward fusion-zone porosity than does ductile iron. The increased susceptibility to porosity is due partly to differences in chemical composition.

Preheating of the base metal is an effective method to prevent cracking of joints in material under 0.75 in. (1.9 cm) thick. Without preheat, the rapid cooling of the heat-affected zone promotes the formation of carbides and martensite. Preheating widens the heat-affected zone and lowers the hardness of the zone by tempering the martensite. To be effective, the preheat temperature should be over 400 F (250 C) and preferably above 600 F (315 C). Highly restrained joints may require preheat temperatures up to 900 F (480 C). Slow cooling of the completed joint is also beneficial because it tempers the martensite and reduces the stress that could result from nonuniform cooling. A blanket of insulation can be placed over the joint to reduce cooling rate.

For joints in material over 0.75 in. (1.9 cm) thick, preheating may not be sufficient to prevent cracking. Peening of each weld bead is an effective technique for joints in heavy sections. Peening with a round-nosed tool deforms the metal and relieves the stresses. Figure 15 shows the effects of peening and preheating on cracking of joints in gray iron.

Cracking of gray iron can also be controlled by careful selection of welding parameters. For example, the heat-affected zone can be widened by the use of a slow travel speed and a high welding current. A wide heat-affected zone is less prone to crack than a narrow zone because of better stress distribution and slower cooling. Also, a multipass welding technique will reduce cracking susceptibility because subsequent beads temper the preceding beads and their heat-affected zones.

Fig. 16 — Groove-joint design used for dissimilar metal joints.



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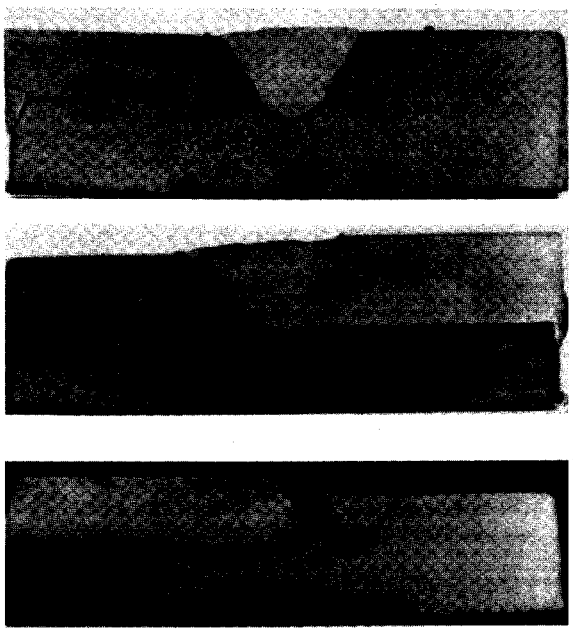


Fig. 17 — Dissimilar joints welded with FC 55. Top: ductile iron to steel. Center: ductile iron to alloy 400. Bottom: ductile iron to Type 310 stainless steel

Dissimilar Joints

Some joints between ductile iron and other materials were welded with FC 55 to investigate the dissimilar-welding capability of the cored electrode. The dissimilar materials were carbon steel, Type 310 stainless steel, and a 67% nickel, 32% copper alloy (MONEL alloy 400).

Ductile iron was joined to the three dissimilar materials by groove joints having the configuration shown in Figure 16. All three combinations were successfully welded. Figure 17 shows cross sections from the joints. The standard procedures for welding ductile iron gave good results for joints involving carbon steel and alloy 400. The joints between ductile iron and stainless steel showed that FC 55 has a low tolerance for chromium dilution. Hot cracking, generally along the centerline of the weld, occurred with excessive chromium dilution. Careful positioning of the electrode and a slow travel speed of 10 in./min (25 cm/min) produced minimum dilution and crack-free welds. Chromium dilution could also be controlled by applying a surfacing weld of ENiFe-CI filler metal on the stainless steel joint surface.

Fillet welds between ductile iron and carbon steel were also evaluated. The joints are illustrated in Figure 18, and the evaluation results are summarized in Table 3. As shown in the table, dilution by the ductile iron is greater than that by the steel, resulting in a longer fillet leg on the ductile-iron side of the joint. If a joint must

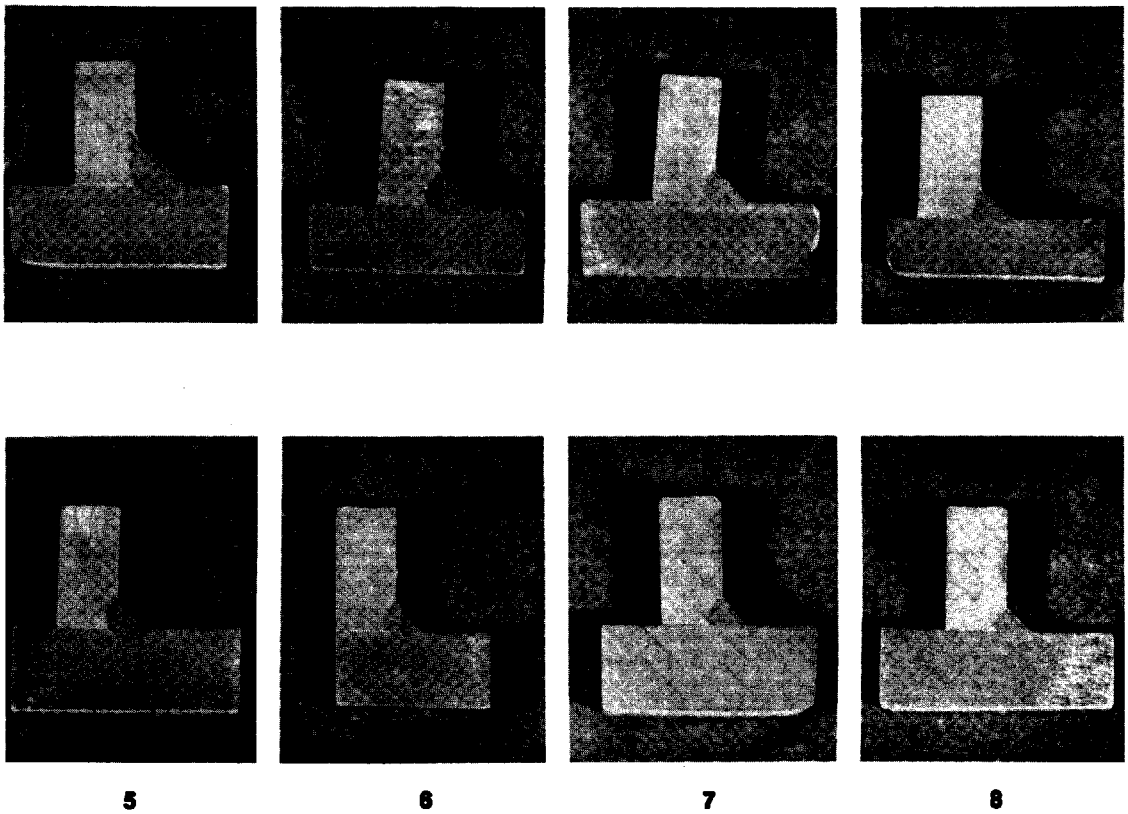


Fig. 18 — Fillet welds between ductile iron (horizontal member) and steel described in Table 3

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Table 3 — Effect of Welding Conditions on Characteristics of Fillet Welds Between Ductile Iron and Steel

No.[a]	Travel speed		Current,	Voltage,	Heat input		HAZ width		Dilution, %		Fillet size,[b] in.		
	in./min	cm/min	A	V	J/in.	J/cm	in.	cm	Cast iron	Steel	x	y	z
1	10	25	350	30	63,000	24,800	0.0137	0.0348	31	10	0.40	0.50	0.65
2	15	38	350	30	42,000	16,500	0.0117	0.0297	29	23	0.30	0.37	0.46
3	20	51	350	30	31,500	12,400	0.0104	0.0264	30	25	0.26	0.28	0.35
4	25	64	450	32	34,500	13,600	0.0107	0.0272	40	17	0.25	0.33	0.38
5	30	76	450	32	28,800	11,300	0.0101	0.0257	36	15	0.24	0.28	0.34
6	35	89	450	33	25,400	10,000	0.0097	0.0246	36	36	0.24	0.24	0.33
7	40	102	450	30	20,250	7,950	0.0095	0.0241	31	36	0.21	0.18	0.29
8	45	114	450	30	18,000	7,100	0.0090	0.0229	30	27	0.19	0.16	0.24

(a) Numbers refer to photographs in Fig. 18.
(b) x = length of fillet leg on steel; y = length of fillet leg on ductile iron; z = bead width.

have equal fillet legs, the electrode should be positioned on the steel side of the joint.

The data in Table 3 also show the effects of variations in heat input on the heat-affected zones of fillet welds. As heat input is lowered, the heat-affected zone becomes more narrow and contains fewer carbides. With a sufficiently low heat input, the heat-affected zone will contain no carbides. Martensite cannot be eliminated by a low heat input, but it can be tempered by a two-pass welding method.

Summary

A new flux-cored electrode has been developed that makes possible, for the first time, high-speed automatic welding of cast irons. The electrode deposits weld metal having a composition similar to that from the widely used ENiFe-CI flux covered electrode for cast irons. The flux-cored electrode can be operated with or without shielding gas and with submerged arc flux. Deposition rates of over 15 lb/hr (6.8 kg/hr) have been obtained. The electrode has produced high-quality joints in both ductile iron and gray iron. It has also been successfully used to join ductile iron to some dissimilar metals.

Acknowledgments

The development and testing of FC 55 and the preparation of this paper involved many people at Huntington Alloys. Special recognition is given J. F. Crickard for his outstanding assistance throughout the project. Among the others who made significant contributions are H. F. Hendershot, G. R. Pease, J. H. Mesick, R. F. Freeman, J. P. Hunt, L. E. Eiselstein, L. E. Shoemaker, M. Morehouse, W. Turko, T. M. Chan, R. E. Avery, and H. R. Conaway.

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The Effect of Base Metal Microstructure on the Impact Properties of Cast Ductile Iron Weldments

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ABSTRACT

Cast ductile iron plates 1.27 cm (0.5 in.) thick were welded with a flux-cored iron-nickel electrode to determine the effect of weld size and base metal microstructure on the amount and continuity of carbides and martensite in the heat-affected zone. The influence of these parameters on the impact properties of the weld was subsequently measured. Large welds (or high heat inputs), high nodule counts and a pearlitic matrix were found to increase the amount of continuous carbides in the partial fusion zone, while small welds (low heat inputs), high nodule counts and a pearlitic matrix increased the incidence of martensite in the weldment. Low impact properties were obtained for small welds, high nodule counts and a pearlitic matrix. These effects can be explained due to the influence of the nodule count and matrix on the rate of solution of graphite in austenite and liquid during the welding cycle.

Introduction

An important limitation of cast ductile iron has been the lack of a superior structural joining method. Poor weldability of ductile iron can be attributed to the formation of massive iron carbides in regions of the parent metal that are partially melted during the weld pass and to the formation of martensite in regions heated above the eutectoid temperature. Both carbide and martensite formation are associated with rapid cooling after the weld pass and contribute to the hard, brittle properties in the partial fusion and heat-affected zones of the weld.

Previous work has demonstrated how these deleterious products form during the welding cycle.[1-3] At high peak temperatures, the graphite nodules begin to dissolve in the austenite, producing a carbon gradient from the nodule into the austenite and resulting in a gradation of transformation products, including martensite, within this region. Partial liquefaction also occurs at higher peak temperatures, giving a liquid that tends to solidify as white cast iron containing massive carbides. The amount of carbides and martensite depends on the peak temperature reached in the weld cycle.

Preheating the ductile iron may help minimize these transformation products. Relatively low preheats are found to reduce or prevent martensite.[4-7] However, moderate preheats often increase the size of the partial fusion zone and thus the amount of massive carbides.[1-3] Only very high preheat temperatures can prevent carbide formation.

Post-heat treatment will temper the martensite or graphitize the massive carbides. However, secondary graphite may form from the martensite during tempering,[8] while chain-like graphite nodules are produced when the carbides are graphitized.[9] Both forms of graphite may reduce the ductility of the weld.

Source: AFS Transactions, 79-73, 1979 (American Foundrymen's Society)

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Table 1. Welding Parameters

Depth of Groove mm	(inch)	Arc Voltage volts	Arc Current amperes	Welding Speed cm/min	(inches/min)	Heat Input cal/cm	(Btu/inch)
3.17	(0.125)	28.0	117.5	18.6	(7.3)	2,536	(26,960)
5.55	(0.219)	31.75	137.5	13.1	(5.2)	4,780	(50,790)
6.35	(0.250)	33.6	156.5	12.7	(5.0)	5,940	(63,100)
7.34	(0.289)	36.65	205.0	10.6	(4.2)	10,160	(108,200)

Oxy-acetylene welding[4,10] and "buttering" techniques[11] have been found to be effective in reducing these hard constituents in the weld.

The effects of heat input (or weld cross-sectional area), the original matrix and the nodule count on the formation of massive carbides and martensite were determined in this investigation. The effects were measured by determining the amount and continuity of the carbides and martensite as well as by impact testing of the welds. The impact test was found to be a useful indicator of the continuity and amount of these brittle constituents.

Procedure

Cast ductile iron plates 1.27 cm (0.5 in.) thick were cast into green sand molds. The base metal was melted and superheated in a high frequency induction furnace to 1510C (2750F), nodulized with 9% MgFeSi and inoculated with 75% FeSi. The iron was poured at 1370C (2500F). The final composition of the iron was 3.40% C, 2.05% Si, 0.28% Mn and 0.025% Mg, with the other elements within usual specifications.

The nodule count was varied by controlling the mold inoculation technique and the cooling rate, using graphite chills. Four levels of nodule count were produced.

Sample 1: 115 nodules/mm^2 obtained with no mold inoculation or chills.

Sample 2: 155 nodules/mm^2 obtained by adding 10 g of crushed inoculant to the gating system of the mold.

Sample 3: 228 nodules/mm^2 obtained by mold inoculation and chilling the bottom surface of the casting with graphite.

Sample 4: 410 nodules/mm^2 obtained by mold inoculation and chilling top and bottom surfaces of the casting with graphite.

The castings were austenitized at 900C (1650F) and either air cooled (normalized) to produce a pearlitic matrix or furnace cooled (annealed) to produce a ferritic matrix. The heat-treated plates were then clamped together in a vise, two at a time, and four different sized holes were drilled between the plates to produce grooves. The depths of the grooves were 3.17 mm (0.125 in.), 5.55 mm (0.219 in.), 6.35 mm (0.25 in.) and 7.34 mm (0.289 in.).

A flux-cored iron-nickel electrode 1.98 mm (0.078 in.) in diameter was used to fill the grooves in each of the plates. The

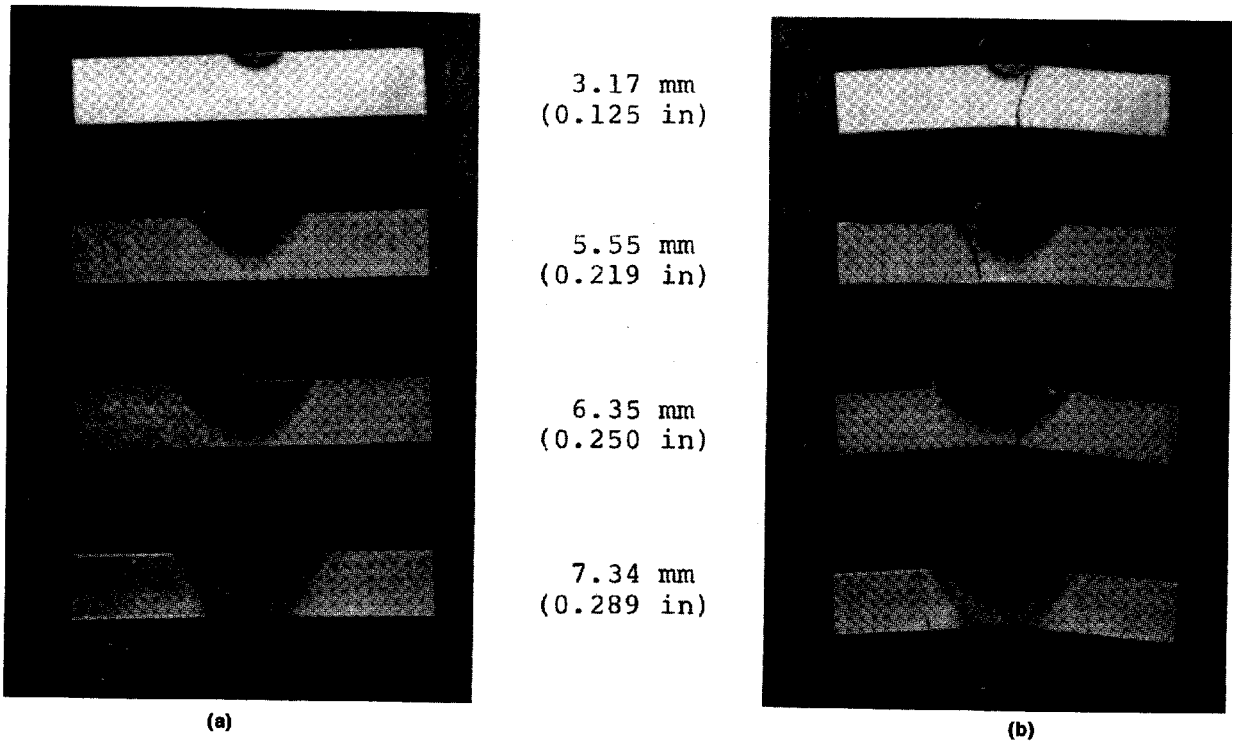


Fig. 1. Typical weld beads from four different groove sizes (a) before and (b) after impact testing. Nital etch.

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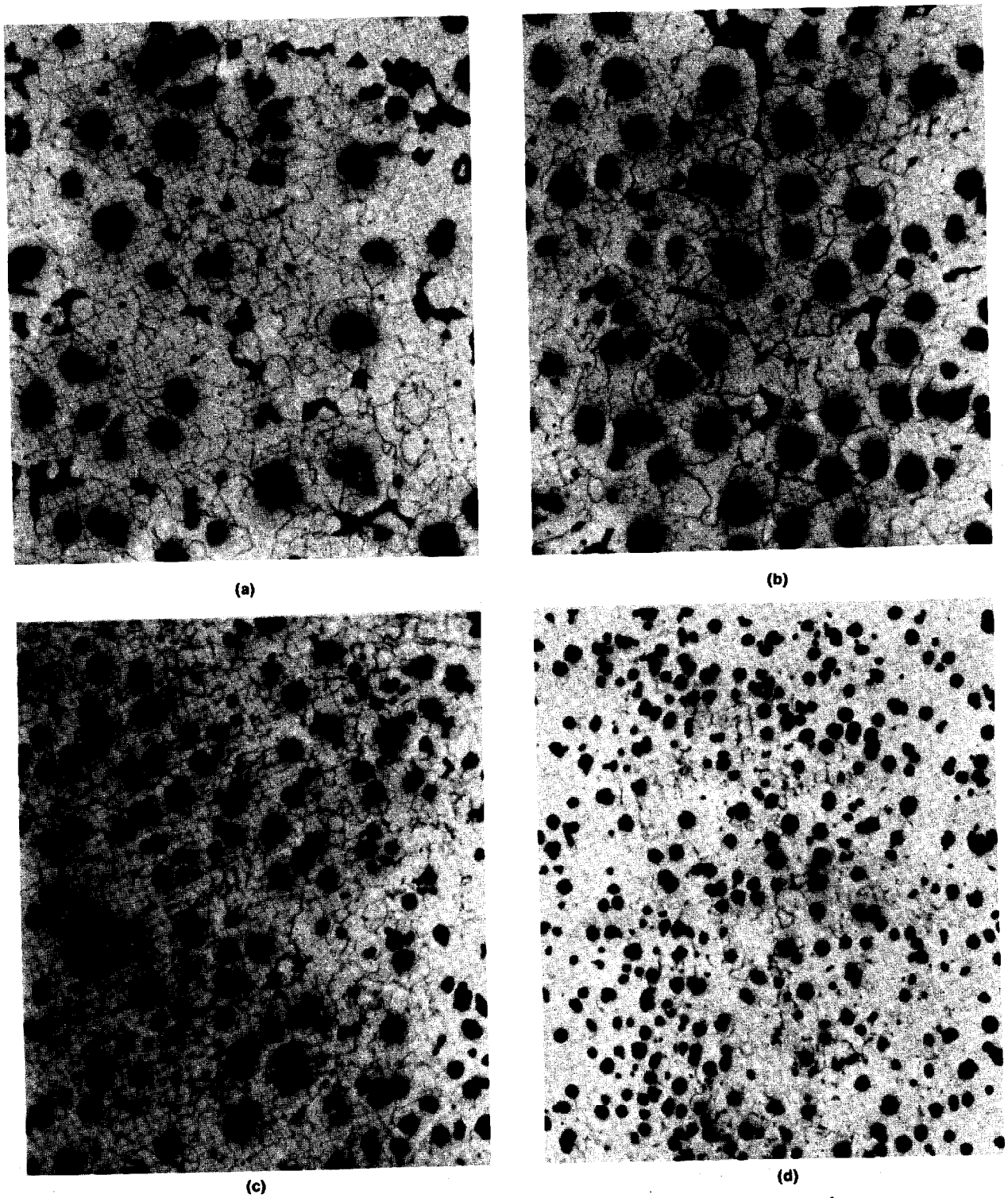


Fig. 2. Typical microstructure of the parent metal after annealing. (a) 115 nodules/mm^2, (b) 155 nodules/mm^2, (c) 228 nodules/mm^2 and (d) 410 nodules/mm^2. Nital etch. 100X

welding parameters and heat inputs are shown in Table 1. The flux-cored iron-nickel electrode was selected because of the good results it reportedly gives in minimizing carbide formation in the weld.[12,13]

Transverse specimens were cut from the welded plates and were surface ground to produce 10 mm x 10 mm x 55 mm unnotched Charpy impact bars, with the weld bead located at the center of the bar. Figure 1 shows typical weld beads from the four groove sizes before and after testing. All bars were tested at room temperature using a 120 lb-ft hammer. Note that failure always started in the partial fusion zone of the weld and went through the carbide or martensite zones.

The specimens were examined metallographically after

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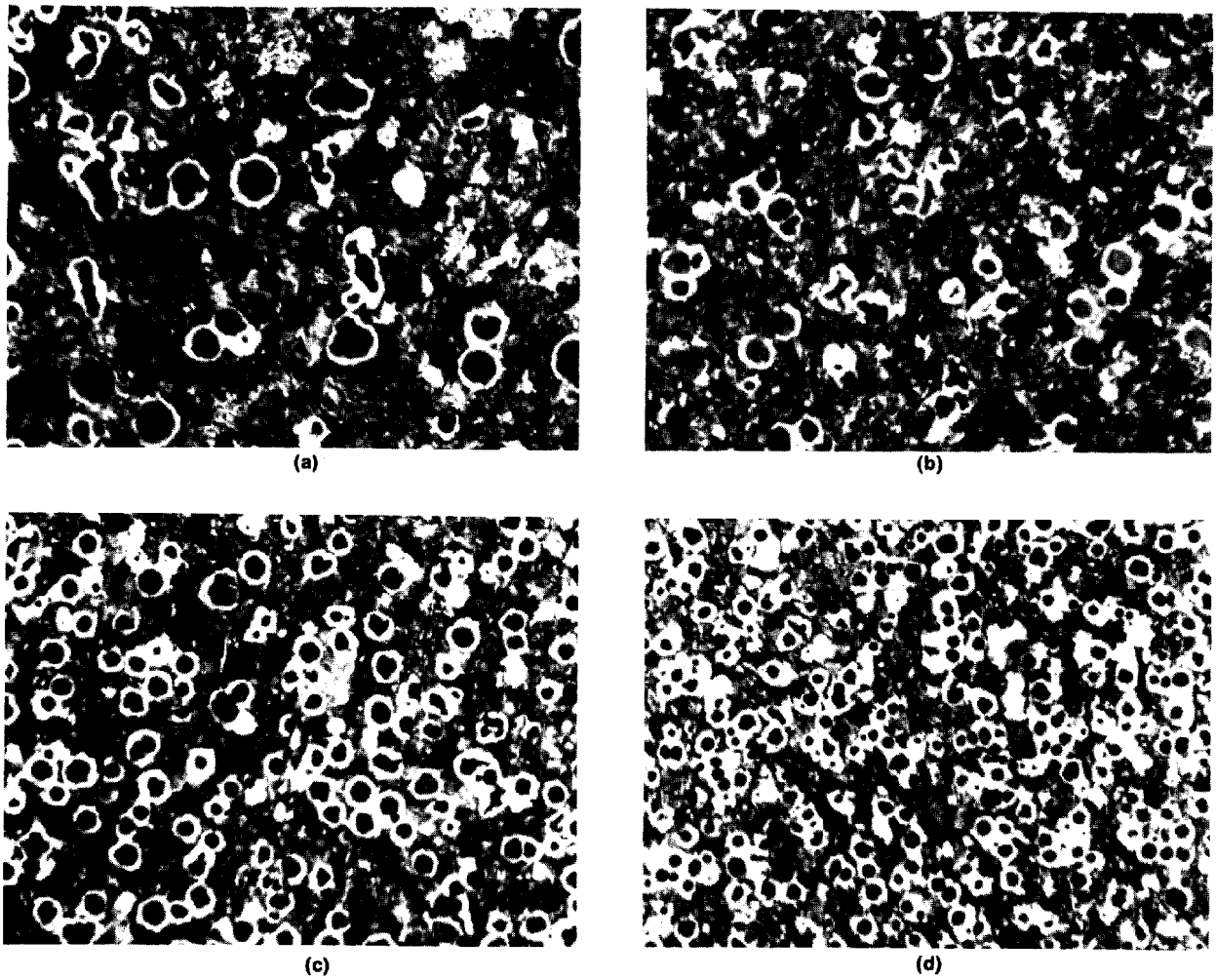


Fig. 3. Typical microstructure of the parent metal after normalizing. (a) 115 nodules/mm^2, (b) 155 nodules/mm^2, (c) 228 nodules/mm^2 and (d) 410 nodules/mm^2. Nital etch. 100X

testing. The nodule count was determined using specimens that had not been etched. The thickness and continuity of the carbide zone was measured after etching the weld with 10% ammonium persulfate, $(NH_4)_2S_2O_8$, which blackens all of the matrix except the carbides and the austenitic fusion zone. Finally, the welds were repolished and etched with 1% nital to reveal the extent of martensite formation.

Results

The original microstructure of the parent metal before welding is shown in Fig. 2 and 3. A relatively wide range of nodule counts was obtained. Annealing was successful in eliminating any intercellular carbides and in producing a predominantly ferritic matrix, although traces of pearlite remained in some of the plates having low nodule counts. Normalizing produced primarily pearlite, except for small ferrite rings around the graphite nodules. The amount of ferrite increased for higher nodule count irons.

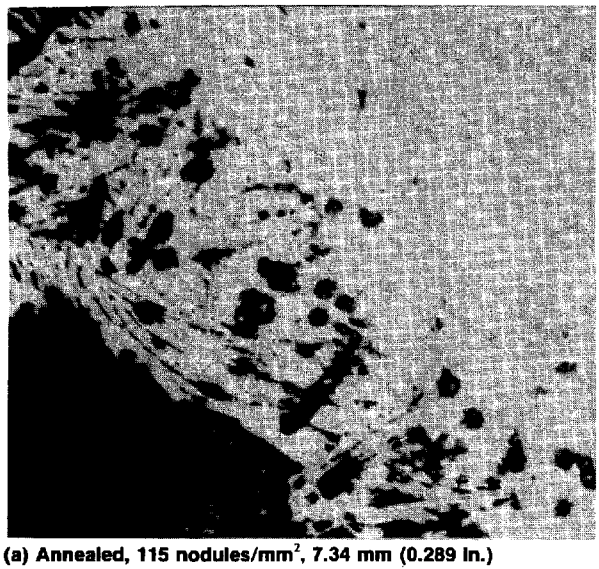
Typical microstructures at the fusion zone boundary are shown in Fig. 4 after etching with the ammonium persulfate solution. The austenitic fusion zone is not attacked, while the heat-affected zone is totally blackened. The massive carbides in the partial fusion zone are easily recognized. It is readily apparent from these examples that the amount, thickness and continuity of the carbides in this region are dramatically influenced by the weld size and base metal microstructure.

The average thickness of the carbide zone is shown in Fig. 5, 6 and 7 as a function of the nodule count and weld size (groove depth) for the annealed and normalized specimens. The carbide zone is increased by increasing the weld size (or heat input) and by increasing the nodule count. These effects can also be seen by comparing Fig. 4a to 4c and Fig. 4a to 4b. At low nodule counts or for small welds, the carbide zone also tends to be discontinuous, as in Fig. 4c and 4d. In general, the pearlitic matrix encourages the formation of carbides in the partial fusion zone and tends to make the zone more continuous, even for small welds (Fig. 4f).

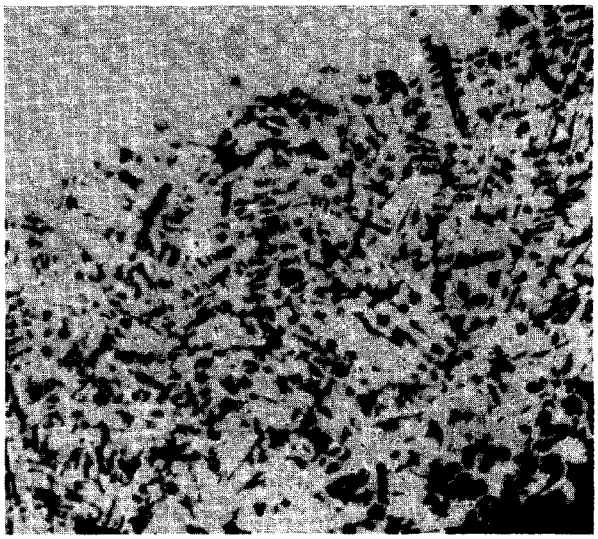
These same regions are shown in Fig. 8 after etching in 1% nital to reveal the structure of the heat-affected zone. Some carbides are now observed in the fusion zone, but the carbides in the partial fusion zone are more difficult to detect. Martensite needles can be observed in the partial fusion zone; again, the amount and distribution of the martensite depends on the weld size and base metal microstructure.

It was not possible to quantitatively measure the effect of these parameters on the formation of martensite. Qualitatively,

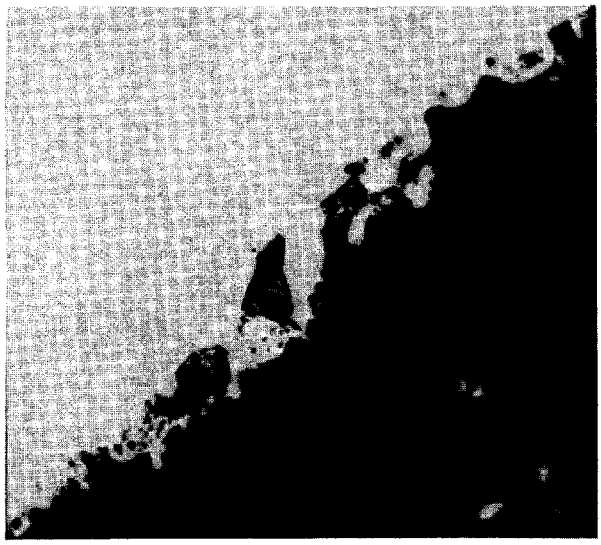
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(a) Annealed, 115 nodules/mm^2, 7.34 mm (0.289 in.)



(b) Annealed, 410 nodules/mm^2, 7.34 mm (0.289 in.)



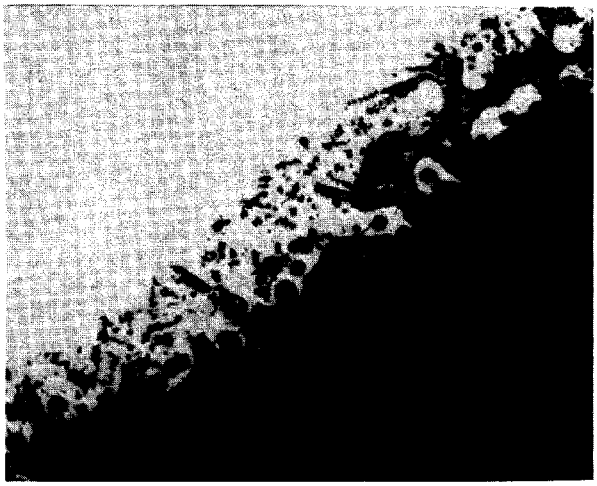
(c) Annealed, 115 nodules/mm^2, 3.17 mm (0.125 in.)



(d) Annealed, 410 nodules/mm^2, 3.17 mm (0.125 in.)



(e) Normalized, 115 nodules/mm^2, 7.34 mm (0.289 in.)



(f) Normalized, 410 nodules/mm^2, 3.17 mm (0.125 in.)

Fig. 4. Typical microstructure of the partial fusion zone showing carbides. Ammonium persulfate etch. 250X

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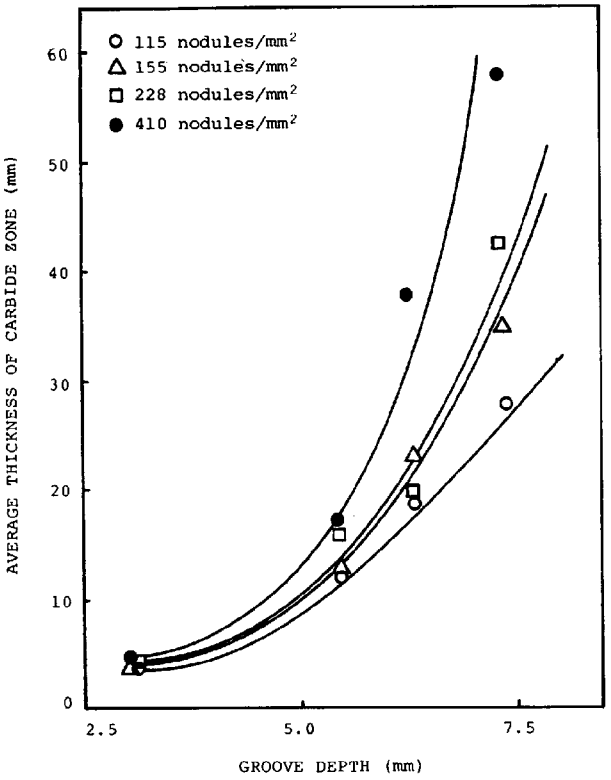


Fig. 5. Average thickness of the carbide zone versus nodule count and weld size for annealed plates.

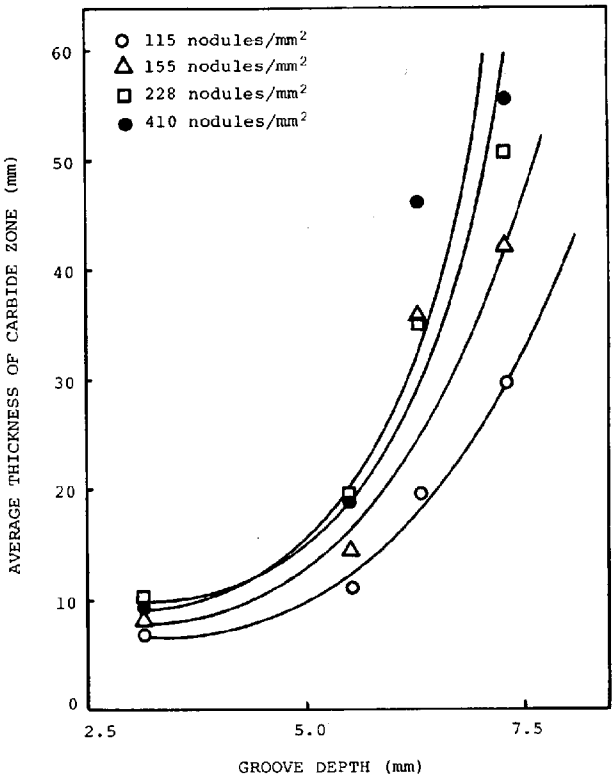


Fig. 6. Average thickness of the carbide zone versus nodule count and weld size for normalized plates.

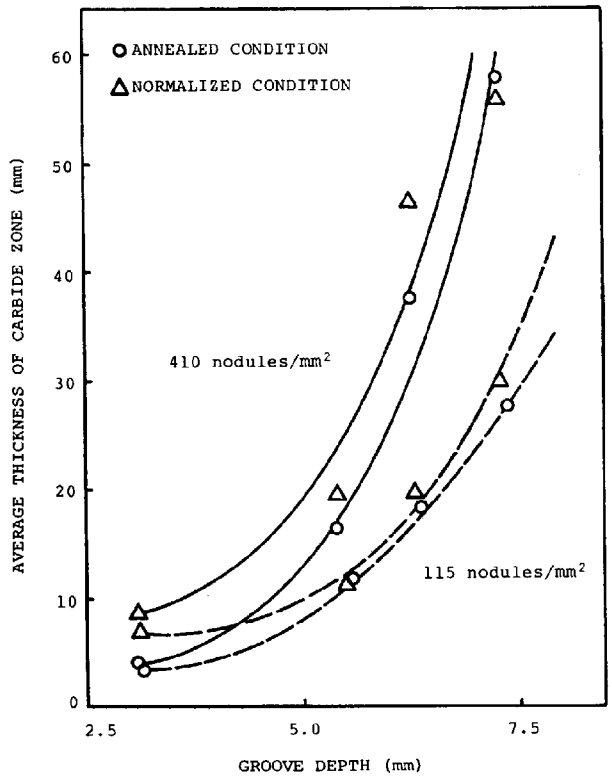


Fig. 7. Average thickness of the carbide zone as affected by the original matrix structure for high and low nodule count irons.

the amount of martensite was found to increase when the weld size decreased (compare Fig. 8a to 8c or Fig. 8b to 8d) and when a pearlitic matrix was welded. Increasing nodule count appears to increase the continuity of the martensite band (compare Fig. 8c to 8d).

Figure 9 shows the effect of nodule count and matrix structure on the impact values for 3.17 mm (0.125 in.) and 7.34 mm (0.289 in.) deep grooves. The impact value is found to decrease with increasing nodule count, with increasing pearlite and with decreasing weld size.

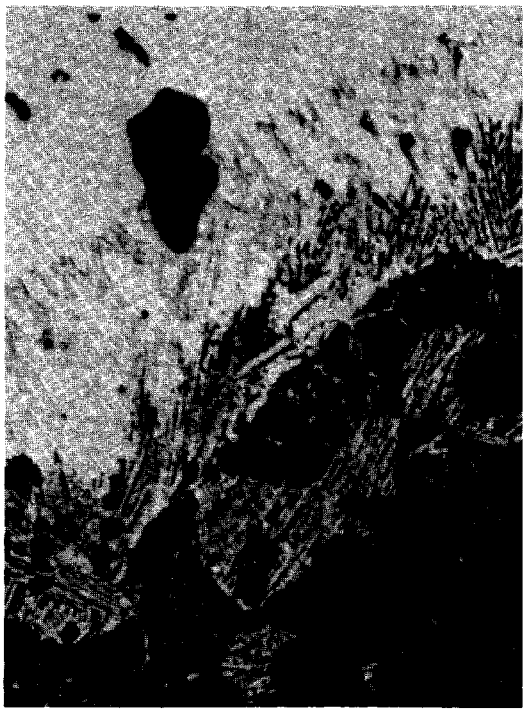
Discussion

The diffusion of carbon from the graphite nodules into the austenite increases with increasing temperature and accelerates rapidly when melting begins. Consequently, a high carbon austenite and a very high carbon liquid can be produced adjacent to the fusion zone.

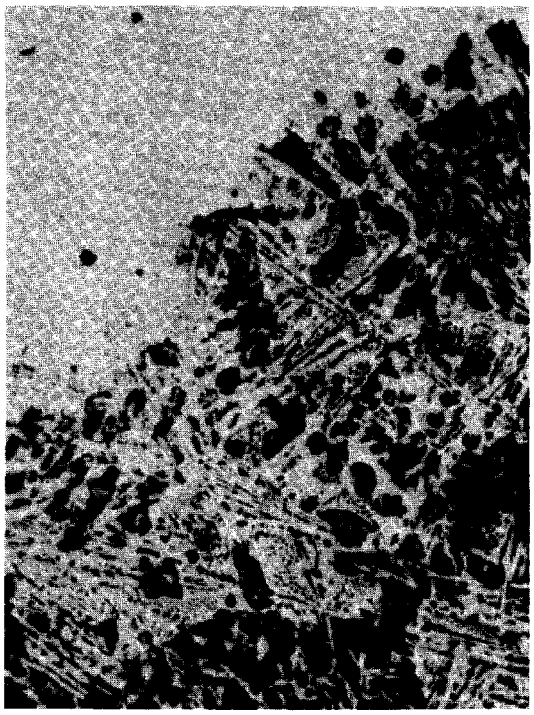
High nodule counts accelerate the enrichment of the austenite and liquid in carbon due to a higher total graphite/austenite interface area and shorter diffusion distances between the graphite nodules. As a result, higher nodule counts would be expected to cause the austenite and liquid to be saturated with carbon more rapidly than low nodule counts, producing more martensite and thicker bands of continuous carbides on cooling, as demonstrated in Fig. 5 to 7. The large amount of carbides then reduces the impact properties of the weld.

More carbides were also found in the larger welds, which require a higher total heat input. The partial fusion zone in the large welds remained at a high temperature for long periods of time, permitting more extensive carbon diffusion into the austenite and liquid. This increased the size of the partial fusion zone and promoted the formation of massive carbides on cooling, as shown in Fig. 5 to 7. Thus, larger welds would be expected to have low impact values. However, the slower

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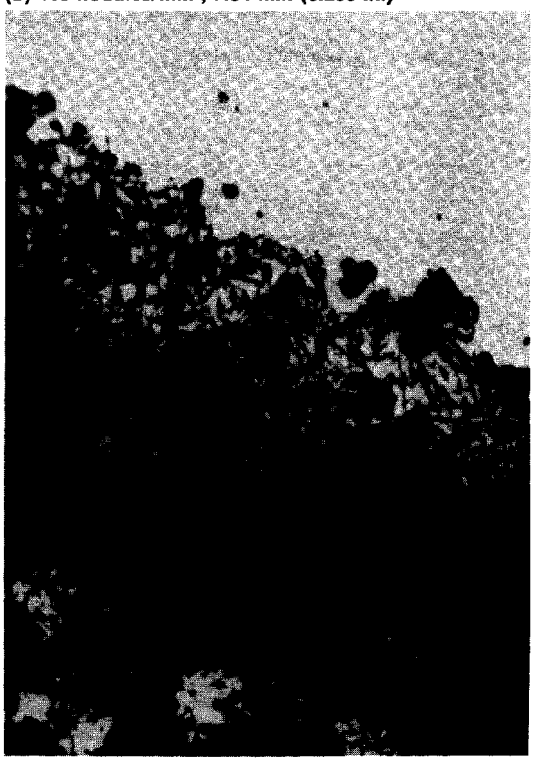
(a) 115 nodules/mm^2, 7.34 mm (0.289 in.)



(b) 410 nodules/mm^2, 7.34 mm (0.289 in.)



(c) 115 nodules/mm^2, 3.17 mm (0.125 in.)



(d) 410 nodules/mm^2, 3.17 mm (0.125 in.)

Fig. 8. Typical microstructure of the partial fusion zone for annealed plates. Nital etch. 250X

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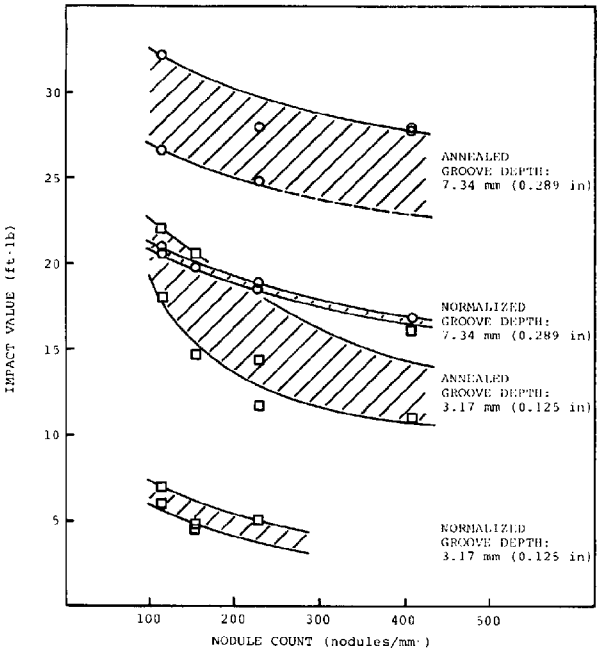


Fig. 9. Effect of nodule count, matrix structure and weld size on impact value.

cooling rates associated with the high heat inputs minimized the formation of martensite. Because the small welds had a continuous band of martensite, the impact values were very low. The martensite appears to be more embrittling than the massive carbides.

Normalized specimens displayed a lower impact strength than annealed specimens. Part of this effect may be due to the base metal itself, because pearlitic irons normally have poorer impact properties than ferritic irons. In addition, the austenite that forms from the dissolution of pearlite initially contains more carbon than the austenite that forms from ferrite. Consequently, a much different carbon gradient is established in the austenite surrounding the graphite nodules. A similar situation exists when the base metal begins to melt. Because of the lower amount of carbon in the austenite and the liquid, less martensite and massive carbides will form in the annealed specimens, as shown in Fig. 7, and the impact properties will be improved. Similar observations have been reported in previous work.[9,14,15]

Summary and Recommendations

The results of this research have demonstrated that the impact properties of ductile iron weldments depend on the amount and distribution of massive carbides and martensite in the heat-affected and partial fusion zones of the weld. In particular, thick continuous bands of massive carbides are detrimental to properties. Martensite may become a problem when the carbide zone is small and discontinuous.

The size of the weld changes the type of undesirable constituent — massive carbides for large welds or martensite for small welds. The smaller welds have the poorer impact properties. However, better overall results might be obtained by minimizing the carbides rather than the martensite, since the martensite can be eliminated at lower post-heat temperatures than the carbides and the products of tempering may be less detrimental than the products of carbide graphitization.

The original microstructure of the parent metal is also significant. Large amounts of pearlite and high nodule counts both favor carbide and martensite formation and thus reduce impact properties.

Control of the microstructure would be beneficial in designing ductile iron castings that are to be fabricated or assembled by welding. The areas to be welded should be slow cooled during solidification of the casting to encourage a relatively low nodule count. Excessive inoculation or chilling would be harmful. The casting should be annealed prior to welding. Heat inputs during welding should be low, assuring that the carbides are discontinuous, but the weld must be tempered to eliminate the severe reduction in impact properties due to martensite formation.

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Submerged Arc Welding of Ductile Iron

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ABSTRACT

The effect of alloying additions in the welding flux on the graphitization and nodulization of the fusion zone in pearlitic ductile iron has been investigated using the submerged arc welding process. Nodulization of the ductile iron fusion zone was achieved using low-carbon steel wire with specific welding flux additions.

Introduction

Ductile iron is a cast material consisting of a steel matrix with spheroidal or nodular graphite. This relatively new engineering material combines the castability of gray cast iron and the mechanical properties of cast steel. Before the full potential of ductile iron can be realized it will be necessary to develop welding procedures[1-17] which enable joints to be completed having properties comparable with those of the base metal.

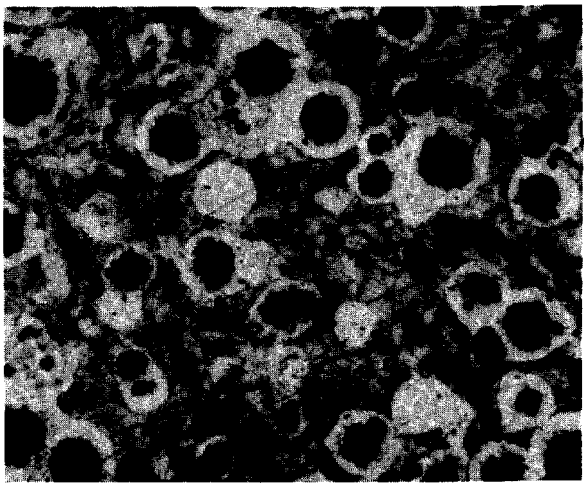
The graphitization and nodulization of the ductile iron weldments is a function of the thermal experience and chemical composition of the weld pool.[18-20] Composition variations of the weld pool result from the dilution of ductile iron parent metal as well as from the melting of the welding wire and flux additions. Graphitization of cast iron weldment is achieved by additions of carbon (C), silicon (Si) and calcium (Ca) to the weld pool. Nodulization of the graphite is promoted by magnesium (Mg) and rare earth (RE) additions to this weld pool. To gain a better understanding of the influences of these flux additions on the microstructure of the ductile iron weldments, this submerged arc welding flux investigation was initiated.

Experimental Procedure

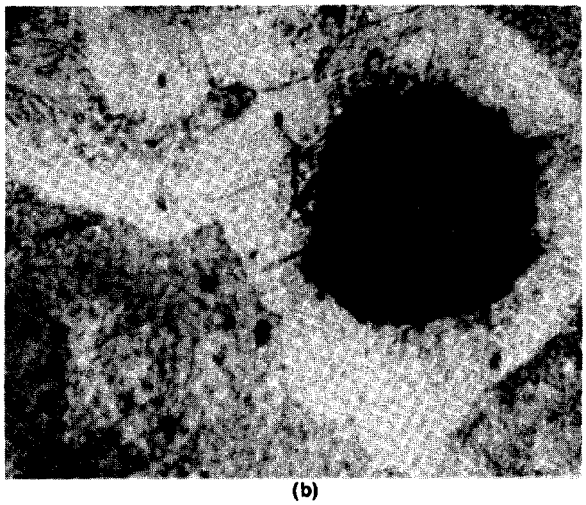
Ductile iron base metal test plates (180 x 80 x 30 mm) were prepared for welding. The chemical composition (wt %) and the physical properties of this ductile iron base metal are shown in Table I. Prior to welding, all welding test plates were ground to remove the oxide scale and the surface was cleaned with alcohol.

The microstructures of the as-received ductile iron plates are shown in Fig. 1a and 1b. The structure consisted of graphite nodules surrounded by ferrite rings and the darker lamellar structure in the matrix is pearlite. This typical structure is described as a bull's-eye pattern. By comparison with standard structures of AFS[21] the base metal consists of 27% ferrite and 73% pearlite. It also has 46 nodules/mm^2 and using the conversion curve for nodule counts of Schwartz,[22] a number of 1650 nodules/mm^3 was obtained. The average size of these nodules in the as-received ductile iron was 0.093 mm in diameter.

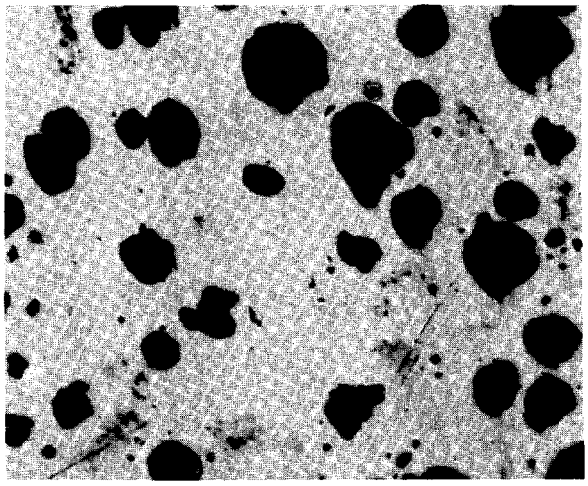
Some of the test plates received a ferritizing heat treatment and their structure is shown in Fig. 1c. It contains graphite nodules and the lighter structure in the matrix is ferrite. Its hardness is 163 Bhn and the graphite nodules have an average diameter of 0.1198 mm. Small irregularities in the shape of the nodules were observed after the ferritizing heat treatment.



(a)



(b)

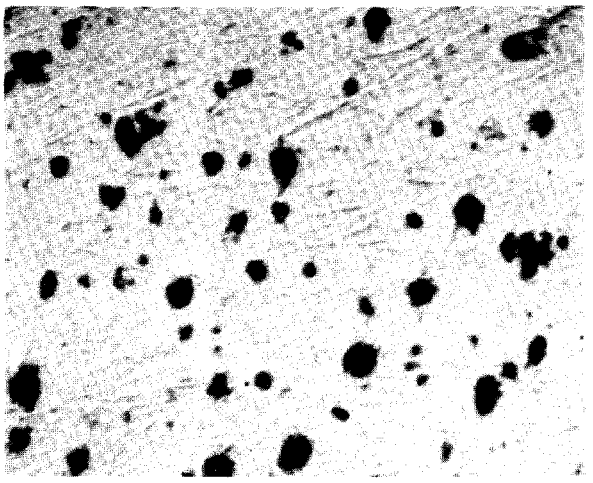


(c)

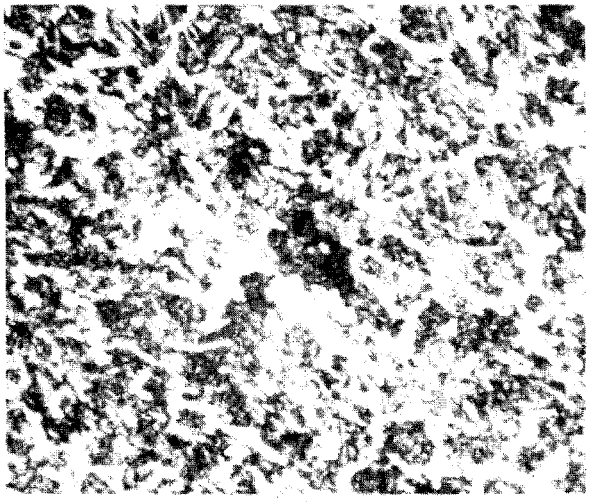
Fig. 1. Microstructure of the ductile iron base metal: a) pearlitic ductile iron (nital etch, 100X), b) micrograph of the graphite nodule surrounded by a ferritic ring in a pearlitic matrix (nital, 800X), c) ferritic ductile iron (ferritizing heat treatment). (nital, 100X).

Source: AFS Transactions, 77-75, 1977 (American Foundrymen's Society)

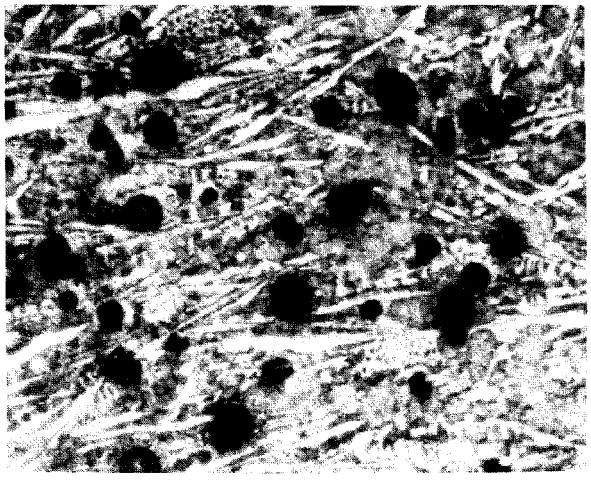
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(a)



(b)



(c)

Fig. 2. Micrograph of the weld metal: a) unetched, 250X, b) ammonium persulphate, 400X, c) etched nital, 250X.

Table 1. Chemical and Physical Properties of the Ductile Iron Base Metal

Composition	Weight Percent
C	3.80
Si	2.64
Mn	0.60
Mg	0.046
Cr	0.092
Mo	0.072
Cu	0.10
P	0.016
Ni	0.08
Tensile strength	92,800 psi
Yield strength	66,000 psi
Elongation	12%
Hardness	217 Bhn

This investigation began using 100% commercial flux. Later, by adding different amounts of graphite, four fluxes were prepared with 2, 5, 10 and 15% graphite with the balance being the commercial flux. An attempt was made to prepare three more fluxes with 20, 25 and 50% graphite but this effort was discontinued because of instability of the welding arc and poor bead morphology.

The research continued by preparing five more fluxes adding 1.5% Si (3% ferrosilicon containing 50% Si) to each composition of the five already prepared fluxes consisting of commercial flux with 0, 2, 5, 10 and 15% graphite additions. In this way ferrosilicon was increased to amounts of 6, 9, 12, 15 and 18%.

Table 2. Chemical Composition (wt%) of Welding Wire

		C	Mn	Si	S	P
Welding wire E1 12 AS.17-69	Specified	0.15 0.17	0.60 0.35	0.05 max.	0.035 max.	0.03 max.
	Check Analysis	0.07	0.54	0.02	0.03	0.009

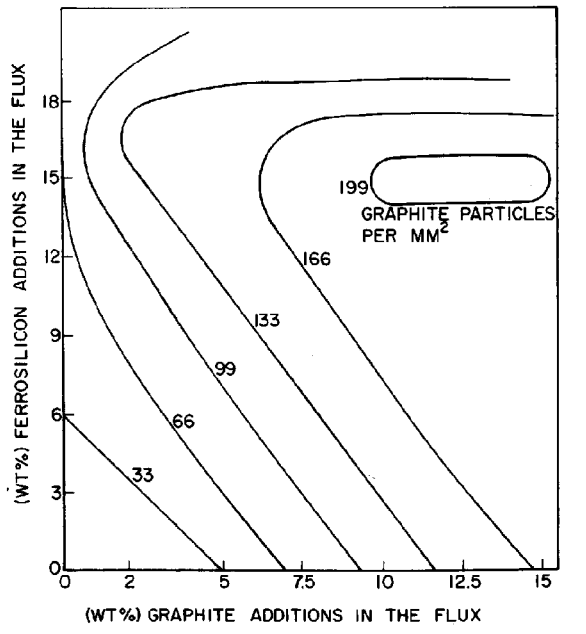


Fig. 3. Constant-graphite-particle lines related to the silicon and graphite additions in the flux.

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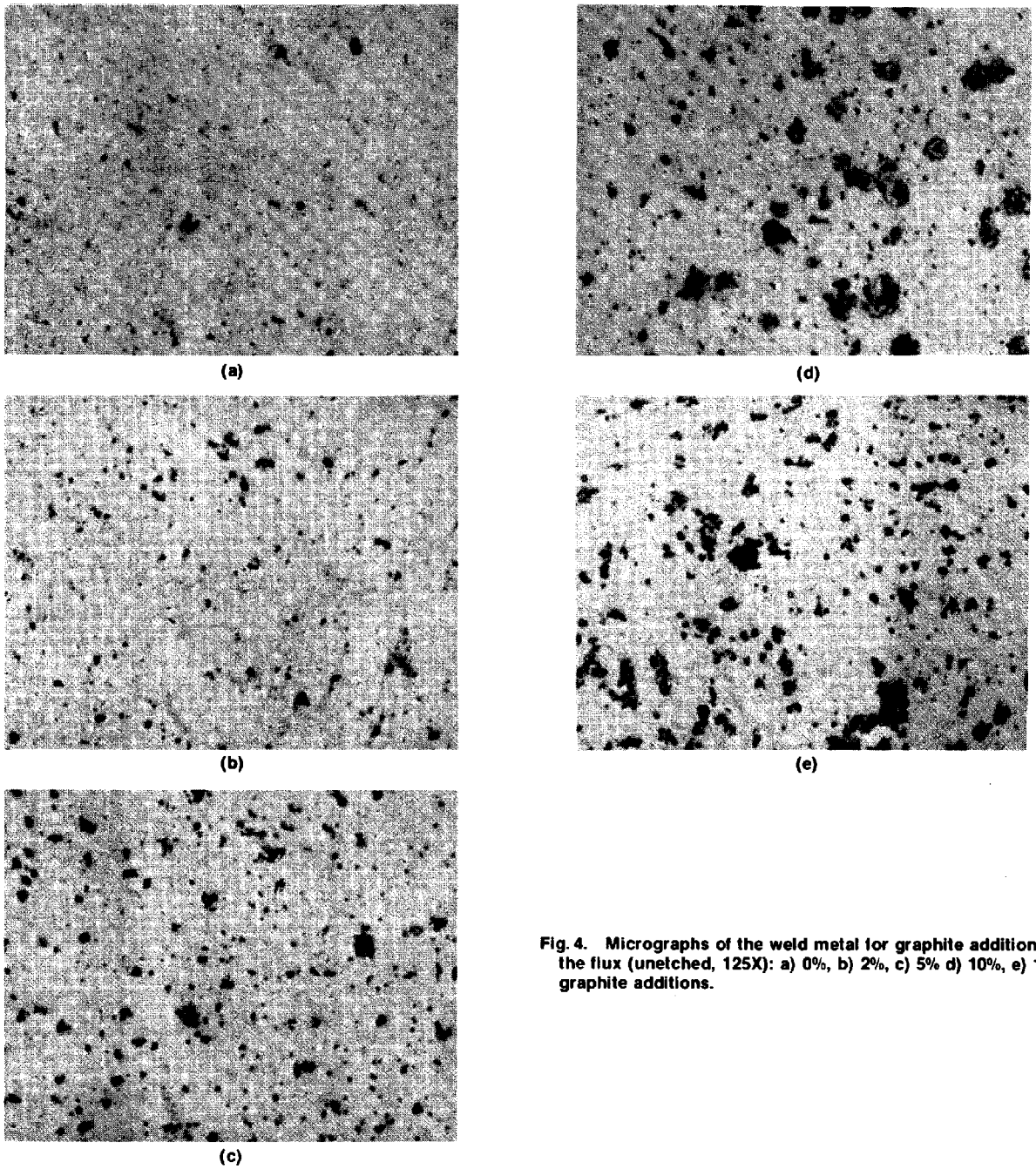


Fig. 4. Micrographs of the weld metal for graphite additions in the flux (unetched, 125X): a) 0%, b) 2%, c) 5% d) 10%, e) 15% graphite additions.

Magnesium in the form of ferromagnesium (3.19% Mg, 46.86% Si) was also used to prepare five more fluxes with 0.04, 0.08, 0.12, 0.16 and 0.20% Mg; 15% graphite; additional ferrosilicon additions to maintain a total of 3% Si in the commercial flux.

Five fluxes were prepared by adding 6% of a low-Si inoculant alloy (47.08% Si; 0.79% Ca; 1.25% Al and the balance Fe) to fluxes containing 0, 2, 5, 10 and 15% graphite additions in the commercial flux. These fluxes were prepared to investigate the influence of Ca additions. A low Mg-containing treatment alloy (3.07% Mg; 2.0% RE and 45.01% Si) in amounts of 0.04, 0.08, 0.12, 0.16 and 0.20% Mg was also used to prepare five more fluxes with 15% graphite, additional ferrosilicon to maintain a total of 3% Si in the commercial flux. These fluxes were prepared to investigate the effect of RE additions. The materials used in making up the fluxes were carefully pulverized, sized and reground until all the material passed through a 28-mesh screen. After weighing each flux addition, the fluxes were mixed to the proper composition for 60 min to guarantee complete homogeneity in the mixture.

The welding was performed under the condition of constant heat input in order to study the effects of flux additions on the graphitization and nodulization effects in the weld pool. All welding was performed with the submerged arc welding process. The filler metal used in this investigation was 1/8-in.-dia welding wire, the composition of which is shown in Table 2. The welding operation was performed using direct current reverse polarity. A bead on plate was welded in a flat position with a single pass. The specimen was allowed to cool to room temperature. The welding parameters employed in this research are summarized in Table 3.

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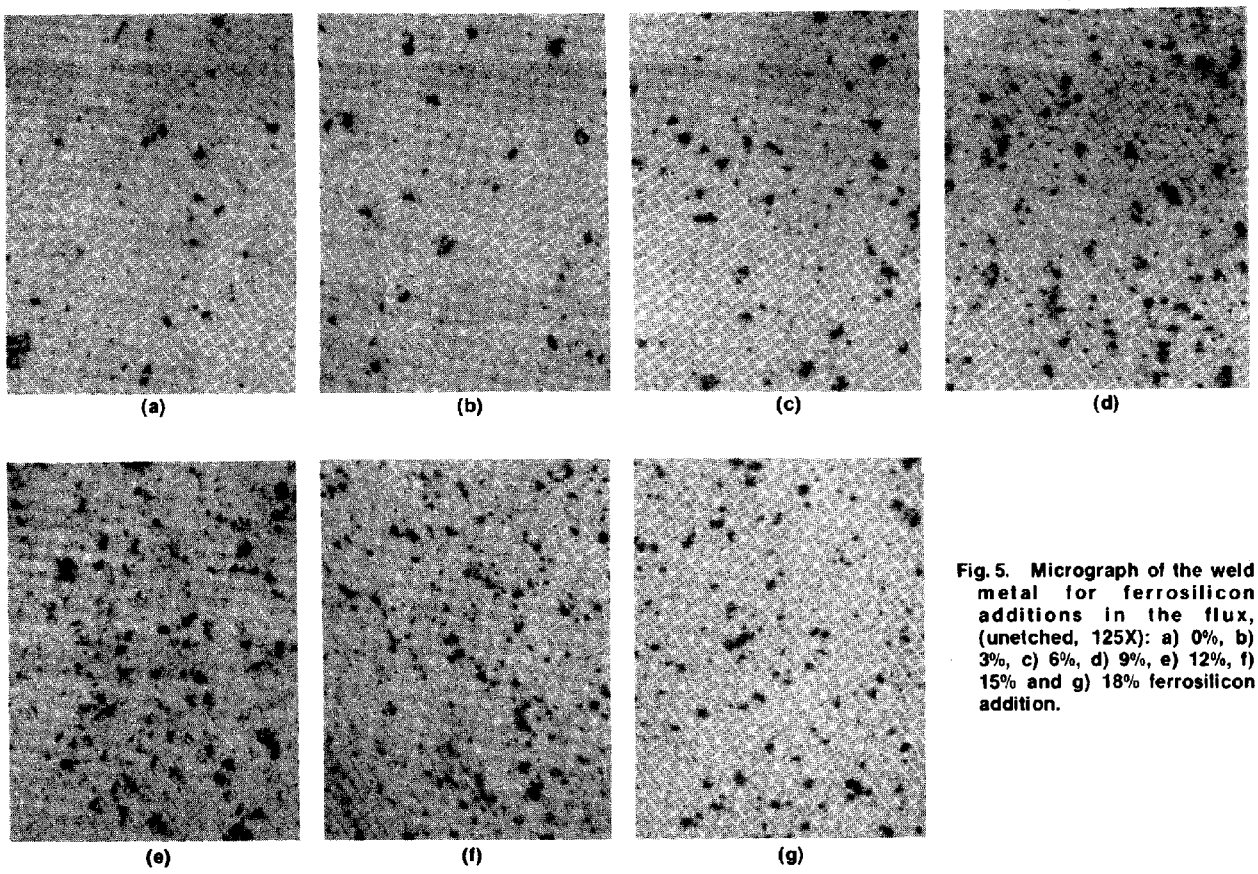


Fig. 5. Micrograph of the weld metal for ferrosilicon additions in the flux, (unetched, 125X): a) 0%, b) 3%, c) 6%, d) 9%, e) 12%, f) 15% and g) 18% ferrosilicon addition.

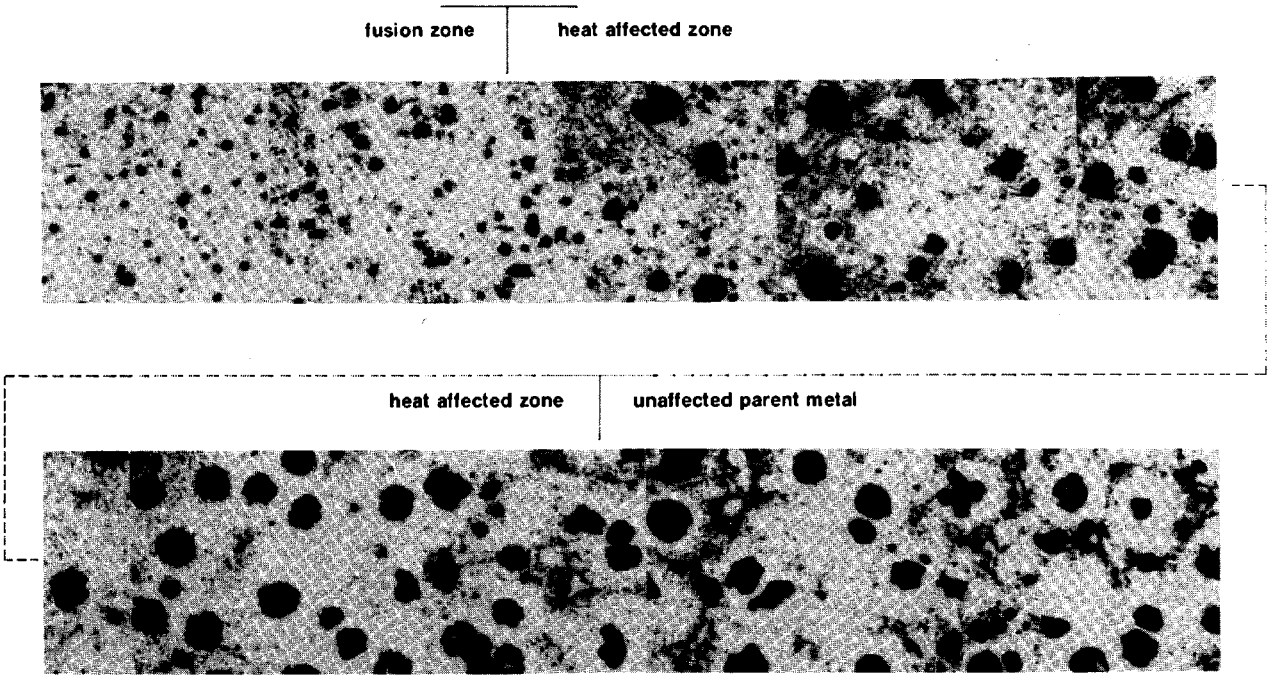


Fig. 6. Sequential microstructure across base metal, heat affected zone, fusion line and fusion zone. (nital 2%, 75X)

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Table 3. Welding Parameters for Ductile Iron Base Metal

Voltage	28
Amperage	300
Travel speed (fpm)	9.6
Wire speed (fpm)	7.2
Heat input (kjl in.)	52.5

Three types of heat treatment were investigated:

1) Preheating of the base metal at 750, 1000, 1250 and 1500F (399, 538, 677 and 816C)
2) Ferritizing heat treatment 3 hr at 1650F (899C), furnace cool to 1275F (690C), hold 5 hr at 1275F (690C), furnace cool to room temperature
3) Tempering: 1 hr at 800, 900, 1000, 1100 and 1200F (427, 482, 538, 593 and 649C), furnace cool. This heat treatment was carried out in a vacuum furnace (0.062 atm).

The specimens, which were obtained from plates welded with each of the different flux compositions, were prepared from a section cut perpendicular to the welded joint. A Rockwell C hardness profile was made from hardness measurements across the weld at 1/8-in. below the plate surface. Several diamond pyramid hardness (dph) tests were carried out to identify carbides, martensite, pearlite and ferrite structures. Optical photomicrographs of the microstructure in the fusion zone of each of the samples were obtained. Measurements of the amount of pearlite and ferrite were accomplished by the outlining and weighing method and by comparing the photomicrograph with standards already established by AFS.[23,24] The carbides in the fusion zone were identified by highlighting with a 10% aqueous solution of ammonium persulfate and comparing with AFS charts.[25] Determination of the number of graphite nodules per unit area was accomplished by using a ground, polished and unetched specimen. Chemical analysis of the specimens was performed to determine the weight percent of Mg, C, Si and Mn in the fusion zone of these test weldments. Two chemical analysis samples were used for each test condition to decrease the uncertainty of these reported values.

Results and Discussion

Effect of the Silicon and Graphite Additions on Flux Behavior

The number and morphology of the graphite particles in the weld metal were observed in unetched weldments as shown in Fig. 2a. When the metal was etched with ammonium persulfate the carbides appear as light particles as shown in Fig. 2b. Also the weld metal was etched with nital 2% to display the complete microstructure as shown in Fig. 2c. About 35 specimens of the weld metal were microscopically examined to observe the influence of the Si and graphite additions to the fluxes. Figure 3 is a graphical correlation of the results relating the graphitization of the weld metal to the Si and the graphite content of the flux. From Fig. 3 it is observed that the graphitization increases with the amount of graphite addition to the flux and this effect is more prominent for the flux with greater Si content. It can be seen that large increases of graphitization in the weld metal require increases in both the Si and graphite content of the flux (Fig. 4 and 5). Generally the graphite in the weld metal was found to be an intermediate shape between spheroidal and flake. This nodular tendency could be due to diluted Mg and RE of the melted base metal. Such behavior was predicted by Riley.[2] The size of graphite in the weld metal goes from very small (0.011 mm) to large (0.044 mm). The amount of carbide increased as the graphite content increased in the flux with no Si content. But for the other flux compositions the amount of carbides in the weld metal remains constant, although the graphitization increased. Loper and Heine[26] reported that an inverse relationship exists between the amount of carbide formation and graphitization.

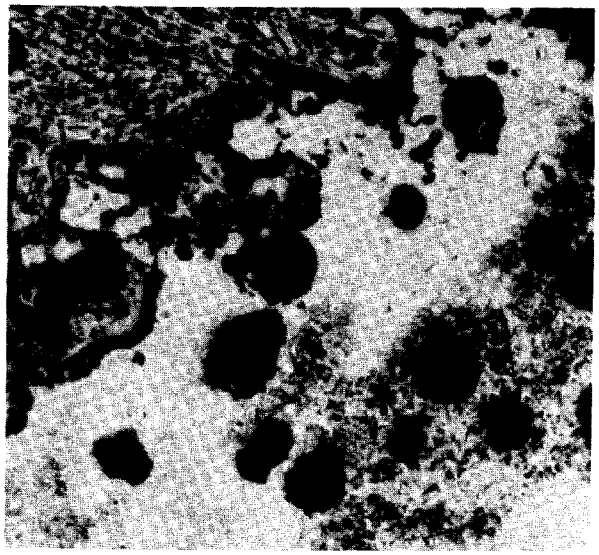


Fig. 7. Microstructure of the fusion line in ductile iron weldment. (2% nital, 100X)

The transition from the pearlitic ductile iron parent metal with the graphite nodules which are surrounded by ferrite rings to the carbide-rich fusion zone is well illustrated in Fig. 6. Notice the partial decomposition of the graphite nodules in the heat affected zone (HAZ) and the resulting martensite. Adjacent to the HAZ is a very narrow region called the fused base metal or the fusion line, Fig. 7. In this region liquid exists around each nodule during welding. After rapid cooling, this liquid region transforms to primary carbides and martensite. The weld structure consisted of intermediate-shape graphite, carbide structures and fine pearlite.

The width to depth ratio (W/D), a parameter typically reported for weldments, is illustrated as a function of the flux addition in Fig. 8. This ratio (W/D) is most sensitive to graphite addition in the flux, while the ferrosilicon addition does not have a major effect on bead morphology. The influence of the flux addition on the bead morphology is probably due to the electrical and thermal properties of these specific fluxes. The flux with the lowest thermal conductivity in the molten state will confine more thermal energy in the location of the weld pool. This will affect penetration and the size and shape of the HAZ.

Hardness profiles for the weldment of pearlitic ductile iron base metal is illustrated in Fig. 9. The high hardness shown (550 Bhn) in the HAZ was obtained at the edge of the fusion zone and

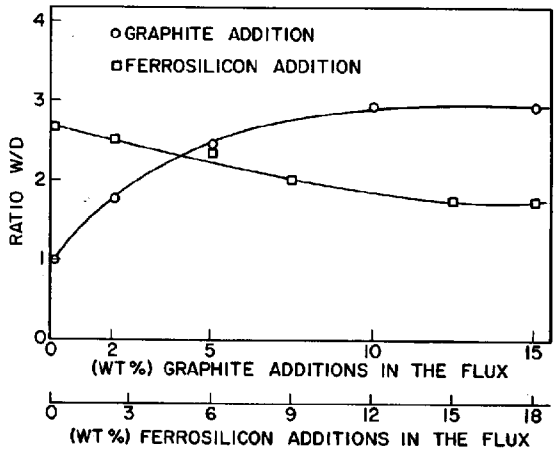


Fig. 8. Ratio W/D of the weldments related to the graphite and ferrosilicon additions in the flux.

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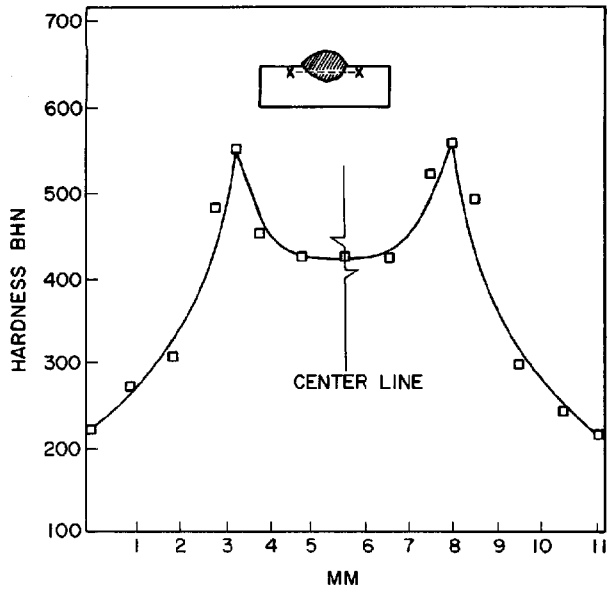


Fig. 9. Transversal hardness profile of the weldments done with a flux of 15% graphite and 6% ferrosilicon addition.

was contained in a very narrow region. It reflects the presence of carbide and martensite at this location. Normally the hardness of the weld metal was 450 to 500 Bhn. This is consistent with the large amount of carbides observed in the weld metal.

After welding 35 specimens with increasing graphite additions of 0-25%, and the ferrosilicon additions of 0-18% in the flux, the selection of the best flux composition was attempted. Fluxes with a composition higher than 15% graphite was eliminated from the selection due to arc instability and inferior bead morphology. Fluxes with a composition higher than 6% ferrosilicon were eliminated due to commercial specifications of Si content in ductile iron of not greater than 3.5% or about the equivalent of 200 Bhn hardness in the ferrite matrix. Ferrite hardnesses greater than 200 Bhn will produce a brittle structure and unmachinable material.

Within the acceptable range of 15% graphite addition and 6% ferrosilicon addition further analysis was necessary to select the best flux composition which produced the most graphite

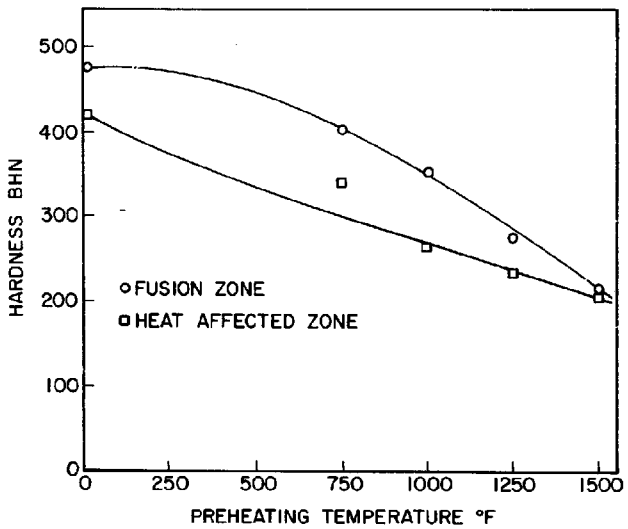


Fig. 11. Hardness profile as function of the preheating temperature of pearlitic ductile iron specimen for the weldments done with 15% graphite and 6% ferrosilicon addition in the flux.

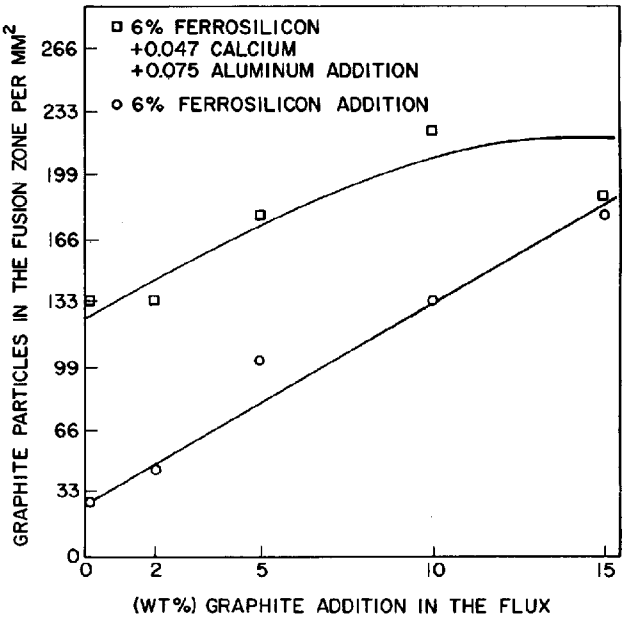


Fig. 10. A comparison of the graphitization effect of calcium and ferrosilicon graphite additions to the flux.

particles with proper graphite morphology in the weld metal (Fig. 3); the optimum flux composition was chosen to be a commercially available submerged arc welding flux with 15% graphite and 6% ferrosilicon additions. It is this flux to which further flux additions (such as ferromagnesium, RE, etc.) were made to continue the investigation of the nodulization effect of the graphite.

Effect of Calcium, Magnesium and Rare Earth Additions on Flux Behavior

Calcium, magnesium and rare earth additions were made to these fluxes (the commercial flux, 15% graphite and 6% ferrosilicon (3% Si)) to investigate their ability to affect the amount and morphology of the graphite. Figure 10 illustrates the effect of Ca, introduced as a ferroaddition, on the concentration of graphite particles in the fusion zone. It can be seen that the small Ca (0.047%) and aluminum (Al) (0.075%) additions to these fluxes of different graphite content caused a major increase in the amount of graphite in the fusion zone. Although excessive additions of Ca have a very powerful

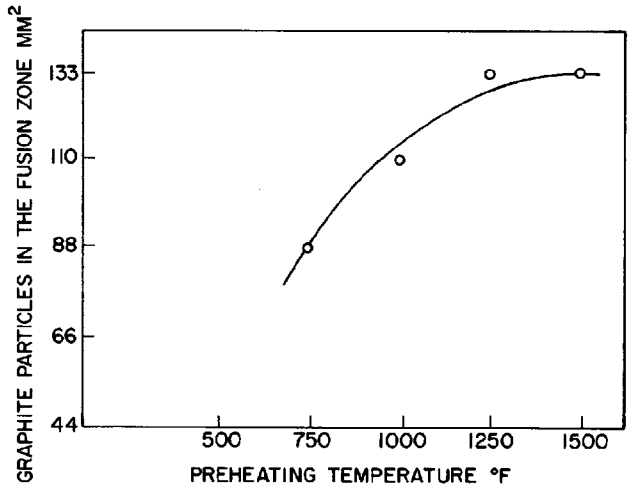


Fig. 12. Graphitization effect in the weld metal related to the preheating temperature of the ductile iron for the weldment done with 15% graphite and 6% ferrosilicon addition in the flux.

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carbide-stabilizing effect, the hardness obtained in the weld metal was about 450 Bhn, remaining fairly constant with respect to the weld metal obtained by using only the ferrosilicon and graphite flux.

Calcium has been found to reduce carbide cell size and increase graphite nodule count.[27] Furthermore it can be observed that larger graphite particle size can be obtained by additions of Ca to the flux. Thus Ca additions would be expected to increase the effectiveness of the inoculants when added in controlled amounts.

A series of fluxes containing a constant 6% ferrosilicon, 15% graphite and increasing percentages of the Mg additions was prepared. It is well known[28] that Mg, in addition to being effective as a deoxidizer and desulfurizer, is used to precipitate graphite as spheroids. Weldment specimens were made with these fluxes for microstructural investigation. Fluxes containing insufficient Mg gave weld deposits which contained flake graphite. The transition from flake graphite to nodular graphite was found in the weld deposits with increasing amounts of ferromagnesium in the flux. It was observed that Mg did not increase the nodule count. The nodule count remained generally constant at about 140 graphite particles/mm^2 in the weld metal. Moreover, a slight increase in the hardness was observed with respect to the ferrosilicon-graphite flux. Vermicular graphite, found in weldments made using ferrosilicon-graphite fluxes, was not found in the weldments made using ferromagnesium fluxes.

Cerium (Ce) was also used in the flux as a RE addition to study the improvement in the spheroidization of the graphite in the weld metal. Five fluxes were prepared with additions of Mg (0.04-0.20%) and Ce (0.024-0.119%). The Si (3%) and graphite additions (15%) remained constant. The Ca additions were reduced from 0.045 to 0.024% in these fluxes. The spheroidal shape was found to start at a lower level of Mg additions in the flux with Ce additions (0.08% Mg - 0.039% Ce). The spheroidal shape of the graphite was obtained in the weld at 0.20% Mg in the fluxes without Ce additions and about 0.08% Mg with Ce additions. There is a definite complimentary effect of the Mg and Ce as spheroidizer elements in the weld metal.

The chemical composition of the weld metal was found to be in general agreement with the microstructure observed in the fusion zone. The number of graphite particles and the C content of the weld metal were found to be a function of the amount of graphite addition to the flux. The Si content of the weld metal and the graphite particle concentration also correlate in the same manner as the ferrosilicon addition in the flux. The efficiency of Si transfer to the weld pool was not found to be a strong function of the amount of ferrosilicon additions in the welding flux. However, lowering the graphite content of the welding flux improves the C transfer efficiency to the weld metal. The chemical analysis for Mg shows a greater transfer efficiency of Mg for lower ferromagnesium additions in the flux as predicted by Clark.[29]

Effect of Preheating and Postheating Treatments on the Microstructure of the Fusion Zone

In addition to the microstructure being controlled by chemical composition, the cooling rate is extremely important. The deposited weld metal cools very rapidly from the molten conditions. This is unavoidable due to the conditions under which it was melted. When the molten ductile iron weld pool had all its C in solution, the casting (parent metal) quickly absorbed the heat from the molten weld metal. Thus the metal, which on ordinary cooling was ductile iron, will produce primarily a carbide structure in the fusion zone. Preheating the ductile iron base metal before welding to slow down the rate of heat loss from the weld metal and reduce the amount of carbide

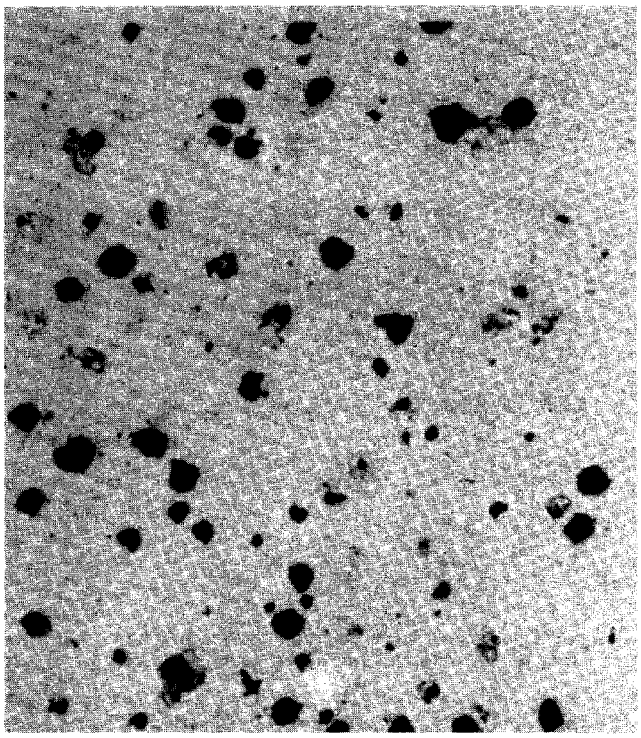


Fig. 13. Microstructure of the weld metal for 6% ferrosilicon, 15%, graphite, 0.20% magnesium additions in the welding flux and a preheating temperature (1200F). (unetched, 150X)

formation is a well-established approach.[9] Figure 11 shows the decrease in hardness of the fusion zone with increasing preheating temperature. Also the concentration of observable graphite particles in the fusion zone was found to increase with increasing preheat temperature (Fig. 12). This increase in the number of graphite particles is consistent with the increase in time for nucleation with the resulting slower rates. The growth of the graphite nodules was enhanced by this slower cooling rate, resulting in increased nodule diameter.

The resulting microstructure of the weld metal which has been preheated at 1200F (649C) for 1 hr can be seen in Fig. 13. This weld was made with the optimum flux additions which were found in this investigation. The flux composition was 6% ferrosilicon, 15% graphite, 0.20% Mg (ferroaddition), 0.12% Ce (ferroaddition) and 0.024% Ca (ferroaddition) to the commercial welding flux. Notice the excellent nodule concentration and shape achieved by the flux and preheat procedure. From the results of this investigation, the above flux composition and preheat treatment have been shown to produce the best nodulization in the fusion zone. This flux composition and preheat treatment are recommended for further research and development as a flux coating for steel electrode in the shielded metal arc welding of pearlitic ductile iron.

Postheat treatment was also investigated as a method to reduce the hardness of the fusion zone and eliminate some of the residual stresses. One hour stress-relieving anneals were applied to the weld metal at several temperatures. The hardness of the fusion zone was found to decrease with increasing post-heat-treating temperature (1-hr anneal).

A full ferritizing anneal (3-hr at 1650F (899C), furnace cool to 1275F (690C), hold 5 hr at 1275F (690C), furnace cool to room temperature) was applied to the weld metal. This anneal changes the hard microstructure (carbide and martensite) in the weld metal into a more stable, uniform and softer microstructure such as ferrite. Four tests at different temperatures and times were carried out. From each of the above tests the time and temperature required to decompose the carbides were recorded

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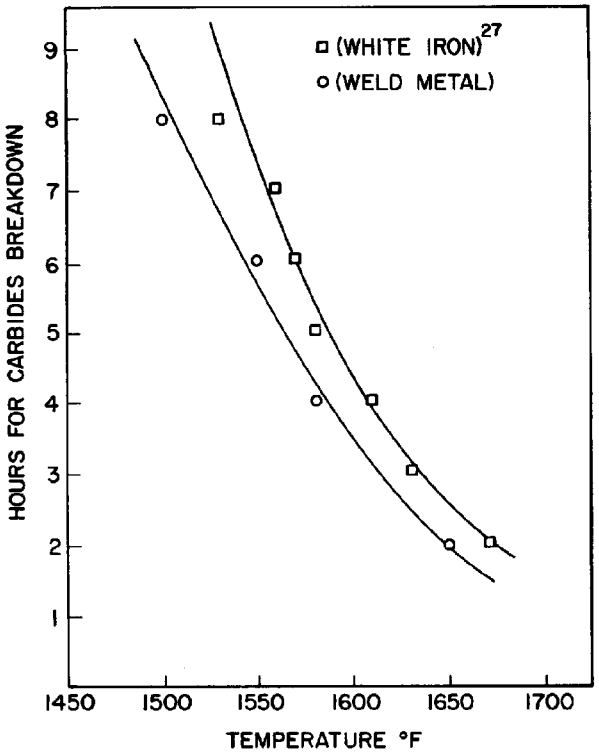


Fig. 14. A comparison of the time required to break down the carbides in the weld metal and the carbides in a standard white iron.

as is shown in Fig. 14. It shows that at any temperature the time required to break down the carbides in the weld metal is less than in white iron. Since the nucleation of the graphite in the weld metal has already been achieved during the welding process by the graphitizing additions to the flux, such a decrease in the necessary time to break down the carbides and to diffuse the carbon to the graphite was expected.

Conclusions

1) The graphitization in the weld metal was increased when Si, Ca and graphite additions were increased in the flux. Also the preheating treatment yielded better graphite size and morphology.
2) The bead morphology was found to be insensitive to variation in the Si additions in the flux. Graphite additions in the flux definitely influence the morphology of the weld metal. Fluxes with greater than 15% graphite addition produced arc instability and poor bead parameters.
3) Neither Mg nor Ce additions to the flux increased the graphite count in the weld metal. However both additions assisted in the precipitation of the graphite in the weld metal as spheres. Magnesium and cerium effects were found to be complementary.
4) Nodulization of the ductile iron fusion zone was achieved using low-carbon steel wire with specific welding flux additions.

Acknowledgements

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Welding Consumable Research For Ductile Iron

Research at the Colorado School of Mines has developed new austenitic consumable materials for welding ductile iron that offer economic advantages, reduce cracking susceptibility, and are machinable.

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Before the full engineering potentials of ductile iron can be realized it will be necessary to develop welding procedures that enable joints to be completed having properties comparable to those of the parent metal. Ductile iron castings could serve as economical substitutes in many steel casting applications if they could be assembled by welding.

There are two approaches to the fusion welding of ductile iron, both of which attempt to control the structures resulting from the eutectic and eutectoid transformations.[1-8] The first approach is to produce nodular graphite during the eutectic reaction by the addition of graphitizing and nodularizing agents to the weld metal from the welding rod, wire, or flux. In this way a weld metal is produced which has a microstructure, mechanical properties, and thermal expansion properties similar to those of the parent metal.

The purpose of the second approach is to produce a tough, relatively soft, austenitic weld metal. Desirable properties of these welds result from avoidance of the eutectoid and its hard products and from the ability of the austenite to absorb carbon rejected by the parent metal, thus reducing the formation of carbides. At present this is accomplished by using nickel alloys as filler materials, such as Fe-95%Ni and Fe-55%Ni. Since there may be possible advantages in the reduction of this high nickel content, the research described in this paper concentrated on studying the effects of the replacement of nickel to varying degrees by manganese in the weld metal.

Experimental Investigation

Fifty 1000g flux samples were prepared by mixing various amounts of 99.9% pure Mn and Ni powder (100 mesh average) with a commercial submerged-arc welding flux. The filler metal used was a 3.2mm diameter low carbon steel welding wire. Weld deposits were produced with the submerged-arc weld process, which was chosen not because its use is advocated for commercial ductile welding, but because as a research tool it offers excellent control of welding variables and a convenient means of changing flux and metal additions to the weld pool. The welding parameters used in this investigation were selected to produce a constant heat input of 3.5kJ/mm. The welding voltage was 30V with a travel speed of 245 mm/min and the current was approximately 500A. Using these parameters and fluxes, single-pass bead-on-plate specimens were made on the pearlitic ductile iron (27% ferrite and 73% pearlite) test plates. The weldments were made in the flat position using DC reverse polarity and without preheat treatment.

Metallographic specimens were prepared from transverse sections of the welds. Fusion zones, fusion lines, and heat-affected zones (HAZ) were examined for microstructures, zone widths, and defects. Volume fractions of austenite phase and carbides in the fusion zone were determined. Figure 1 illustrates the various modifications in zone microstructures resulting from the nickel and manganese additions. Note the breaking up of the continuous carbide structure with these alloy additions. Figure 2 shows the volume percent of carbides in the fusion zone as a function of flux composition. With no alloy additions to the flux the structure is carbides and very fine pearlite, but the maximum amount of carbides in the fusion zone resulted with a

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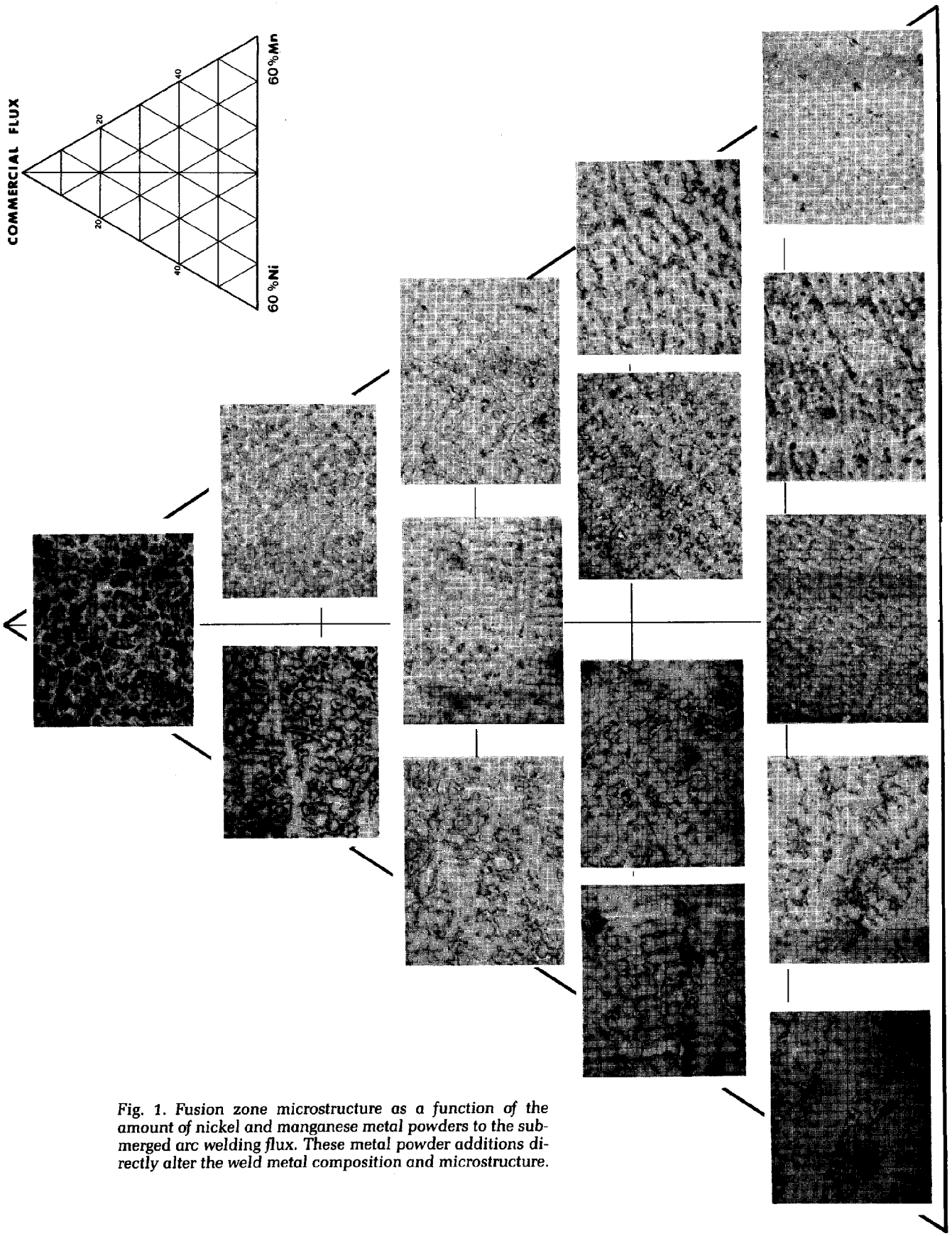


Fig. 1. Fusion zone microstructure as a function of the amount of nickel and manganese metal powders to the submerged arc welding flux. These metal powder additions directly alter the weld metal composition and microstructure.

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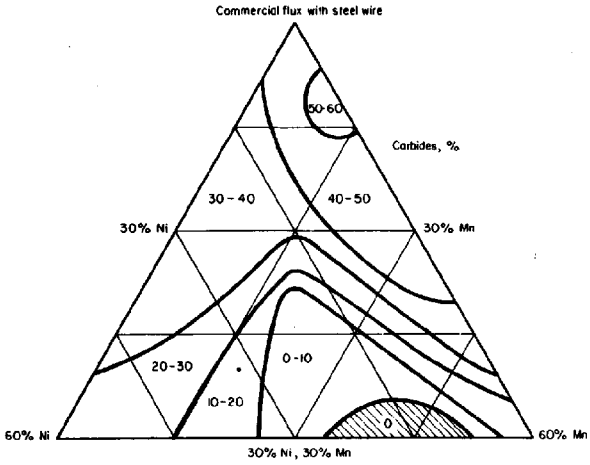


Fig. 2. *Volume percent carbide in weld metal as a function of nickel and manganese powder additions in the welding flux.*

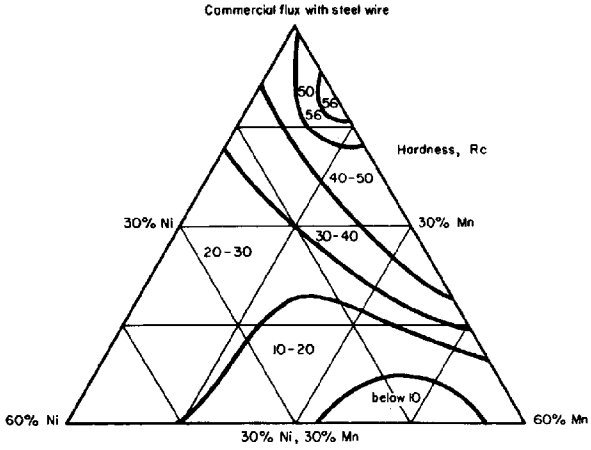


Fig. 3. *Weld metal hardness as a function of the nickel and manganese powder additions in the welding flux.*

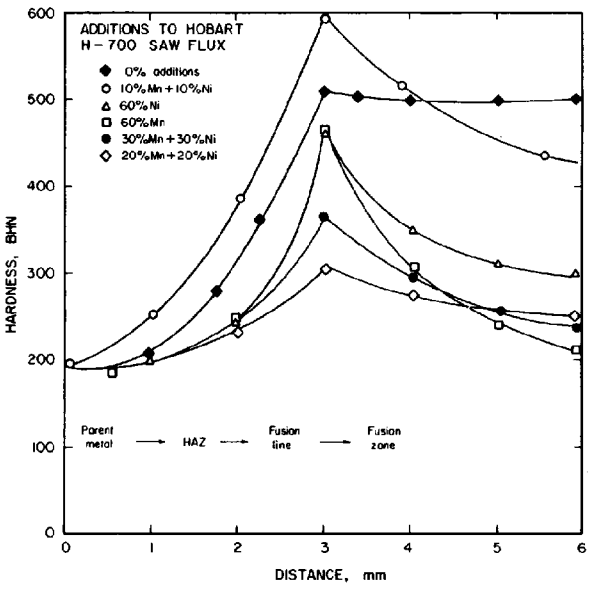


Fig. 4. *Hardness profile across the fusion line of the weldments as a function of nickel and manganese powder additions in the welding flux.*

10%Mn addition to the flux. Small amounts of nickel in the flux are more effective than small amounts of manganese in reducing the amount of carbides, but for large additions this trend is reversed, as indicated by the close spacing of the lines on the lower part of the manganese side of this diagram.

Manganese additions are found not only more effective than nickel in reducing amounts of carbides, but are also more effective in making the carbides discontinuous when comparable amounts of carbide are present. Carbide-free fusion zones, produced with fluxes containing both manganese and nickel, and their composition range, are shown cross-hatched in Fig. 2.

Fluxes that produced fusion zones with discontinuous carbides occupy a larger range, lying roughly within the 0-10% carbide region. Drilling tests have demonstrated that carbide continuity is as serious a concern to machinability as the carbide content. Since hardness is strongly influenced by microstructure, the fusion zone hardness profiles in Fig. 3 are very similar to carbide distribution as seen in Fig. 2. Notice that the region that lacked carbides had a hardness value below 10 Rockwell C.

The hardness profiles in Fig. 4 show large changes in peak fusion line hardness with variations in flux composition. Where temperatures are high enough in the heat affected zone, the graphite nodules partially decompose to produce a shell of high-carbon liquid around the nodules. With the high cooling rates, a closed carbide shell surrounding each nodule results.

Bordering on the fusion zone, the carbide shells may

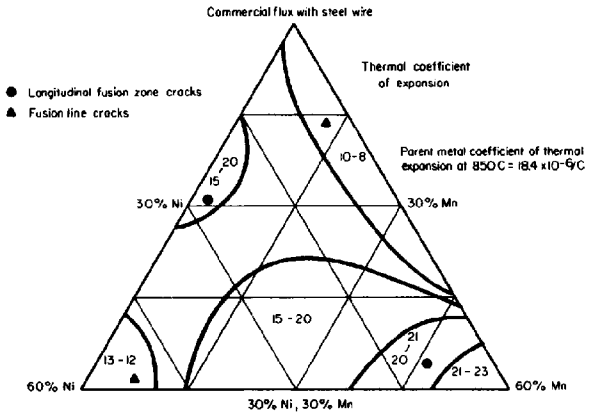


Fig. 5. *Average coefficient of linear thermal expansion of weld metal made with various nickel and manganese powder additions to the welding flux at 850C (1562F), 10⁻⁶/C.*

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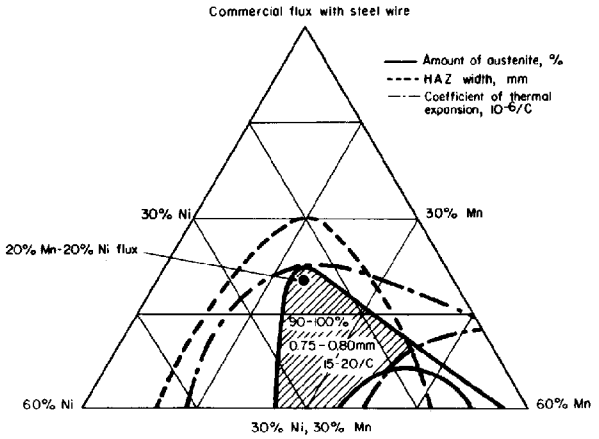


Fig. 6. Range of acceptable flux compositions to produce desirable microstructures, HAZ widths, and coefficients of linear thermal expansion in weldments of ductile iron.

connect or partially connect to each other to form a hard fusion line region. Continuous carbide formation along the partially melted region of the heat affected zone introduces poor mechanical integrity to the weldment. The lack of carbides in the fusion line found for certain compositions suggests that some of these fusion zones are of compositions which melt below the parent metal eutectic temperature. The concern over continuous carbide formation in the heat affected zone also suggests using welding procedures with low heat inputs.

Thermal Expansion

Since thermal expansion mismatch between parent and weld metals is a major cause of cracking, the coefficient of thermal expansion was measured using dilatometry for selected weld samples, and linear thermal expansion coefficients were calculated. The effects of flux composition on thermal expansion are presented in Fig. 5, which shows the value of the coefficent of thermal expansion at 850C (1562F).

As shown, the 60%Ni addition to the flux produces welds with lower thermal expansion than that of the parent metal, but welds made with the 60%Mn flux have larger coefficients of thermal expansion. Also shown are symbols that indicate the types of cracking found in each zone. When the coefficient of thermal expansion is larger than that of the ductile iron parent metal, it appears to be susceptible to fusion zone cracks, and when less, fusion line cracks appear to be promoted.

It is apparent that to reduce the cracking tendencies the coefficient of thermal expansion of the weld metal should match that of the parent metal. Suitable values of the coefficient of thermal expansion and acceptable hardness can be produced with certain fluxes along the 60% total alloy addition line, but satisfactory results may also be obtained, using smaller alloy additions, by changing the flux composition to one near the 20%Mn-20%Ni addition to the flux as shown in Fig. 6.

Optimum Composition

The determination of an optimum fusion zone composition was based on considerations of cracking susceptibility, machinability, and expected economy. Cracking problems can best be avoided by producing welds which are free of continuous fusion line carbides, soft fusion zones with discontinuous carbides, and coefficients of thermal expansion which match that of the ductile iron parent metal. Machinability is also good when the fusion lines are soft and when the fusion zone has small amounts of discontinuous carbides in a soft austenite matrix. The best economy is achieved by reducing total alloy content and by substituting manganese for nickel as long as doing so does not compromise soundness and machinability of the weldment.

Analysis of the fusion zone microstructure, HAZ width, and thermal expansion have shown that properties which will produce the least cracking and best machinability will be found in welds with at least 90% austenite, the narrowest possible HAZ (0.75-0.80 mm), and a thermal expansion coefficient of about 18.4×10^{-6}/C at 850C. The ranges of flux compositions that produce each of these desirable properties are shown in Fig. 5. The cross-hatched area common to these separate ranges is the range of metal additions to the flux that should produce optimum weld properties. When economy is considered, the 20%Mn-20%Ni addition to the flux is probably the best, though higher alloy compositions could be useful in some specific applications, such as when dilution is very high.

Chemical analysis of the weld beads was also performed to calculate transferability of alloy additions from the flux and to estimate the composition of shielded metal-arc welding electrodes and gas metal-arc welding wire that would produce the same results. Chemical analysis of the fusion zone showed that the 20%Mn-20%Ni addition to the flux produces a fusion zone with 13.1%Mn and 15.9%Ni. The wire composition needed to produce an equivalent weld will vary according to the amount of dilution by the parent metal. For 30% dilution, which is reasonable for SMA and GMA welding processes, the ferrous wire would have to contain 18.7%Mn and 22.7%Ni. This is a substantial reduction in nickel content from the 50 to 55%Ni rod and wire commonly used at present.

Further Research

Welding consumable research for ductile iron is continuing at the Colorado School of Mines. New investigations to determine the influence of copper additions to reduce nickel content of the austenitic weld metal are in progress. These copper additions should also reduce the necessary heat input and assist in alleviating the problem of continuous carbide formation along the fusion line. Also, the influence of the welding parameters on the ductile iron weld microstructures and properties is being studied.

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Joining of Ductile Iron by Several Arc Welding Methods

Metal arc and inert arc welding of magnesium-containing cast iron, using nickel-iron electrodes, steel electrodes and nickel-iron filler wire

by T. E. Kihlgren and H. C. Waugh

INTRODUCTION

"DUCTILE IRON" is a relatively new engineering material which combines the castability and machinability of gray cast iron and the mechanical properties of cast steel to a remarkable degree. The material, sometimes referred to as "spheroidal-graphite cast iron," or "magnesium-containing cast iron" is characterized by a structure in which graphite is present as spheroids instead of the flake form found in gray cast iron. It is this structural feature which accounts largely for the unique characteristics of spheroidal graphite cast iron.

In many of the fields of application of spheroidal graphite cast iron, weldability is an important consideration. With the increasingly wide acceptance of Ductile Iron, the need for more information on its response to welding as a means of fabrication, repair and salvage has become progressively greater.

This paper deals with metal arc welding and to a lesser degree with inert arc welding of magnesium-containing cast iron, using nickel-iron electrodes and filler wire. Brief data are also presented on metal arc welding of this material with steel electrodes.

COMPOSITION AND PROPERTIES OF UNWELDED DUCTILE IRON

Some comment may be in order on the composition and properties of magnesium-containing cast irons. Basically, Ductile Iron is a gray cast iron which has been treated with magnesium, the residual magnesium preferably being of the order of 0.05–0.08%. The base composition can be further strengthened by increases in manganese and nickel content. The following tabulation represents a satisfactory range of compositions of Ductile Iron, unwelded:

C	3.2/4.2
Si	1.0/4.0
Mn	0.1/0.8
P, max	0.10
Ni	0/3.5
Mg	0.05/0.10

The mechanical properties, depending upon whether the casting is in the as-cast or annealed condition will range as shown in Table 1. Using certain compositions and appropriate heat treatments involving quenching and drawing, tensile strengths as high as 150,000 psi with 9% elongation have been obtained.

Table 1

Tensile strength, psi	60,000–105,000
Yield strength, psi	45,000– 75,000
Elongation, %	1–20
Bhn	140–290

The greater portion of current production of Ductile Iron castings would fall within the following range of composition and properties shown in Table 2. Most castings are used in either the as-cast or annealed condition, the anneal normally consisting of 1 hr at 1650° F followed by slow cooling to 1275° F, holding for 5 hr and air cooling. In the paper, the above condition will be designated as annealed or FHT.

Table 2

Composition range	*Properties*
3.2–3.8 C	60/95,000 psi tensile strength
1.75–2.75 Si	45/65,000 psi yield strength
0.20–0.80 Mn	3/20% elongation
0–1.0 Ni	150/250 Bhn
0.05–0.08 Mg	
0.10 P, max	
0.02 S, max	

Most of the data to be presented will refer to base plate falling within the range of composition of Table 1. Generally, low-alloy plate containing less than 0.35 Mn and 0.75 Ni was used for the welding tests.

Part I. Metal Arc Welding

METAL ARC WELDING WITH NICKEL-IRON ELECTRODES

A considerable amount of experience has been obtained on welding of Ductile Iron with coated austenitic nickel-iron electrodes. The particular electrode used for these studies is designated commercially as Ni-Rod 55 and is essentially a 60/40 nickel-iron core wire, coated with a carboniferous lime-spar type of flux coating. There are several other commercially available nickel-iron electrodes differing primarily in the composition of the flux coating applied. Much of the following recommendations on welding procedure will apply to these electrodes also. Since the coating is an important consideration, it is obvious that the data on weld cracking, mechanical properties, etc., cannot be applied indiscriminately to all electrodes of this general description. In the succeeding portions of the paper, when the term "nickel-iron" electrode is used, the data will be based on Ni-Rod 55 experience. In using other commercial electrodes of this type, the electrode manufacturer's instructions with respect to electrode manipulation, welding currents and welding procedure should be consulted.

CHOICE OF CASTING COMPOSITION FOR WELDING

In general, the higher the manganese

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and nickel contents (and hence the weld hardenability) the less latitude there is in welding and the greater will be the need for preheat. For example, a Ductile Iron containing 0.90 Mn and 1.5 Ni will require more care in welding than one with 0.25 Mn and 0.70 Ni, other things being the same. The increase in weld hardenability of the former will be reflected in greater machining difficulties in the heat-affected zone. It is also important to control phosphorus content, if weld cracking is to be avoided. A 0.10% P maximum appears satisfactory when plates are welded with nickel-iron electrodes or filler wire. It is believed that a 0.05% maximum may provide further latitude and is probably necessary where mild steel electrodes are to be used. In general, in applications where weldability is a prime consideration, the composition range of Table 2 should be further narrowed with respect to manganese, nickel and possibly phosphorus along the lines indicated above. The following nominal composition, for example, would be expected to show a good combination of mechanical properties and weldability using several welding processes: 3.50 C, 2.40 Si, 0.25 Mn, 0.70 Ni, <0.05 P, 0.06 Mg.

EFFECT OF PLATE CONDITION

For maximum welding response, Ductile Iron castings should be welded in the annealed (FHT) condition. This treatment insures the greatest freedom from fusion line cracking hazards, as well as the best machining qualities, at least as far as as-welded joints are concerned. Good quality joints can be obtained on either as-cast or annealed material using nickel-iron electrodes or filler wire, but when using mild steel electrodes, the prior anneal is necessary, as will be discussed later.

ARC-WELDING PROCEDURE WITH NICKEL-IRON ELECTRODES

The following procedure is suggested, much of it being applicable to welding with other electrodes and processes.

Preparation of Castings

This phase of any welding job is important, and successful welding may hinge upon the care taken at this stage. On new castings, the preparation consists mainly of removing the casting skin and adhering foreign matter from the surfaces to be joined and from the immediately adjacent area. This can be done by machining, grinding, shot blasting or other suitable means. Such a precaution not only decreases danger of contamination but also insures better slag coverage and removal characteristics.

Castings which have been in service and require weld repairs may be found to be oil impregnated. Such castings are degreased by use of a solvent, such as carbon tetrachloride, sometimes in conjunction with cleaning by steam, which will serve to remove surface contamination with oil and grease. In some instances, oil may have penetrated the casting sufficiently so that torch heating to about 750° F may be indicated to decompose oil and grease and drive off volatile matter.

Welding Amperages

The nickel-iron electrodes under discussion operate on dc or ac. When used on dc, reverse polarity is recommended. The welding current should be as low as possible, consistent with smooth operation, good bead contour and the obtaining of a good "wash." Recommended ranges are given in Table 3.

Table 3

Electrode	*DC*	*AC*
3/32	50–80	60–90
1/8	90–110	100–120
5/32	120–140	130–150
3/16	140–160	150–170

Electrode Manipulation and Welding Sequence

In general, the electrode should be operated with a normal weave, and conventional width beads deposited. The stringer bead technique often used in cast-iron welding is not recommended with the nickel-iron electrode. In making butt joints, a 60- to 75-deg vee should be employed. Where a considerable amount of metal must be deposited, as in filling in of a large cavity, the sides may be "buttered" and the cavity filled in gradually toward the center of the repair area. In any event, the welding should be staggered so that the heat is distributed reasonably uniformly through the weldment.

Peening

This is often done to reduce the stresses in the joint. If peening is to be done, it should be by repeated moderate blows rather than by a few heavy blows, peening each bead as deposited. A 1/2–3/4-in. diam rounded peening tool or a ball-peen hammer is suitable. Peening, properly done, can be quite effective in reduction of stresses. A 6- x 6- x 1/2-in. joint, unpeened and unrestrained, exhibits considerable bowing. The same joint, suitably peened, can be completed with virtually no bowing.

Preheating

In no instance should welding be done on cold castings, as might occur in unheated areas in cold weather, and a preheat of at least 150–200° F should be applied under these conditions. Low-alloy Ductile Iron in the annealed condition may be welded without preheat in many instances. The higher the alloy content (and hardenability) or the greater the restraint, the greater is the need for preheat. For a given composition, as-cast material will be more apt to need preheat than annealed material. In any event a preheat is never harmful, and is a good precaution to observe in order to minimize danger of fusion line cracking. Obviously the amount of metal to be deposited and the severity of subsequent service conditions of the weldment must be considered in deciding upon the use and the amount of preheat. While a preheat of 300–350° F is often adequate, one of 550–600° F is preferred where high stresses are to be expected. An added advantage of preheat is the greater facility of obtaining a good bead contour and a good wash.

Continuous Welding vs. Intermittent

In welding of cast iron, intermittent welding is frequently used, a method in which 2 or 3 in. of weld are deposited, the bead allowed to cool until just warm to the touch, another short bead deposited, and so on. Using nickel-iron electrodes, this procedure is not usually employed in welding either cast iron or Ductile Iron. Continuous welding is used, as in normal welding practice, the heat input of the early passes serving to supply a moderate preheat. However, the bead sequence should still call for a staggering procedure to get reasonably uniform distribution of heat in all parts of the weldment.

Position Operability

The particular nickel-iron electrode used is operable in vertical and overhead as well as in downhand positions. Figures 1–3 show 1/2- x 6- x 6-in. butt welds in all positions, partially completed so that the contour of intermediate passes may be seen. When operating out of position, a smaller electrode and lower amperages should be used than for flat welding, as is customarily the case in welding other metals. For example, in Figs 2 and 3 a 1/8-in. electrode at about 105 amp was used, while the downhand weld was made with 5/32-in. electrode at 135–140 amp. A 1/8-in. electrode, as used in a down-hand weld on 1/2-in. plate, would be operated at about 110–115 amp.

PROPERTIES OF NICKEL-IRON ARC-WELDED JOINTS

The data shown below are averages of two groups of tests, based on standard 0.505-in. and subsize 0.357-in. round tensiles machined from butt welds in 1-in. and 1/2-in. plate, respectively. Welds

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Fig. 1 Flat butt weld in annealed 1/2-in. Ductile Iron plate. Welded with nickel-iron electrode

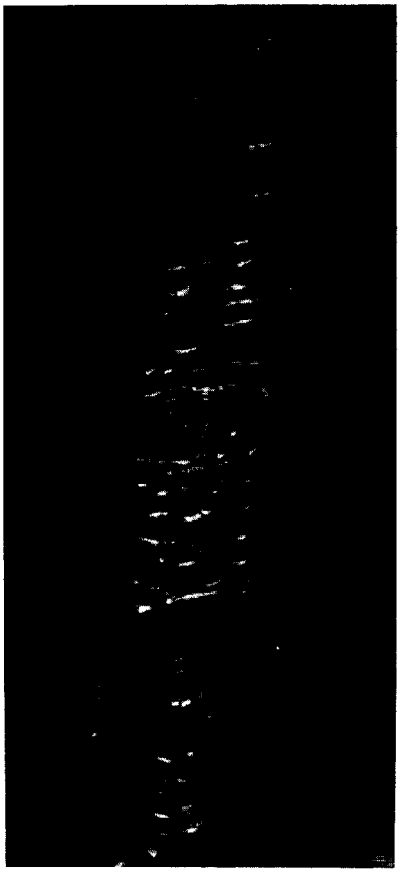


Fig. 2 Vertical butt weld in 1/2-in. plate of annealed Ductile Iron using the nickel-iron electrode

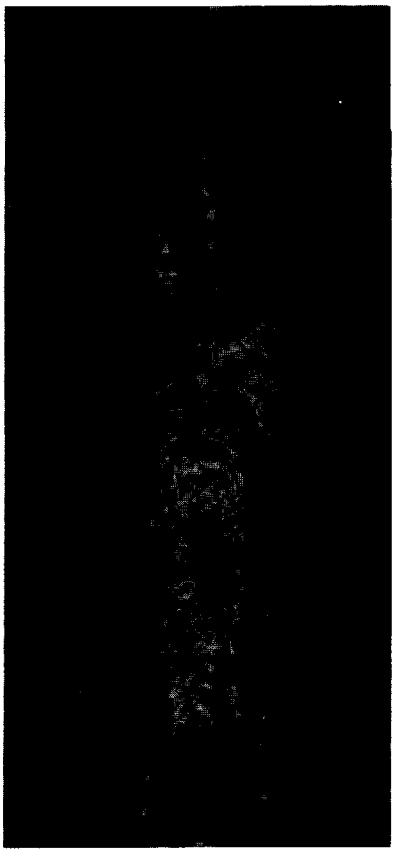


Fig. 3 Overhead butt weld in 1/2-in. as-cast Ductile Iron plate using the nickel-iron electrode

without preheat and with a 600° F preheat are grouped together since no important differences were consistently present. The base plate contained about 3.30 C, 2.30 Si, 0.70 Ni, 0.35 Mn, 0.05 Mg, <0.05 P. Welds were made with 5/32-in. diam nickel-iron electrodes at about 130–140 amp, d-c, reverse polarity. In the case of the welds in 1/2-in. plate, the welded joints were 6 x 6 x 1/2 in. and in the 1-in. plate, 5 x 10 x 1 in. A double-vee, 70-deg included angle, a 1/8-in. root gap and a copper backing were used. The welds were quite severely restrained with four C clamps and were unpeened.

The data illustrate the value of welding the plate in the annealed (FHT) condition, such joints exhibiting fairly good ductility without a postweld heat treatment. The effect of the postweld full anneal on Ductile Iron welded in the as-cast condition is of course marked, with respect to ductility. A 1275° F treatment applied directly to the as-welded joint without the prior 1650° F treatment was also helpful and developed similar properties to the FHT specimens in the one set of tests made.

Figure 4 is a photograph of a set of tensile bars showing location of fracture and the properties obtained on the particular specimens illustrated.

Table 4—Average Mechanical Properties of Nickel-Iron Arc-Welded Joints in Ductile Iron

Plate condition	*Postweld, HT**	*Yield strength, psi, 0.2% offset*	*Tensile strength, psi*	*Elongation in 2 in.*
As-cast	AW	56,900	65,000	1.5
As-cast	FHT	45,900	63,800	10
As-cast	FT	47,700	61,600	5
Annealed (FHT)	AW	44,500	60,600	9
Annealed (FHT)	FHT	42,500	60,500	12
Annealed (FHT)	FT	44,700	60,700	13

* AW = as = welded; FHT = 1650F-1-FC + 1275-5-AC; FT = 1275F-5-AC.

HARDNESS OF NICKEL-IRON ARC-WELDED JOINTS

The nickel-iron electrode deposits weld metal having a Vhn of about 160, in the absence of plate dilution. The as-welded hardness of the weld metal, when diluted with the ductile iron base metal, is well above this figure and varies considerably, depending on the extent of dilution. In general, the weld metal of a multilayer single-vee butt joint in 1/2-in. ductile iron plate will exhibit about 225 Vhn in the as-welded condition.

In single-layer welds, the weld hardness can be kept at a satisfactorily low level simply by use of moderately heavy beads. For example, a light bead deposited upon a cast-iron plate, using a rather rapid speed of travel, may show hardnesses of the order of 350–380 Vhn due to excessive plate dilution. By depositing a substantially heavier (but not abnormally so) bead at the same amperage, using a slower welding speed, the increased ratio of electrode metal to base metal results in the much lower hardness range of 180–225 Vhn. Similar effects are experienced in ductile iron welding.

In operations where single-bead welds are to be made, the use of the so-called heavy-pass technique can be quite helpful where moderately low weld metal hardnesses are

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desired. In multilayer welding, the technique is not required for the above reason, but it has been found useful, especially in the root passes of the weld, in providing further insurance against weld hot cracking in welding heavy sections and is quite generally employed.

Hardness traverses were made on a set of Ductile Iron joints produced with nickel-iron electrode on a Ductile Iron of the following composition: 3.67 TC, 2.35 Si, 0.23 Mn, 0.73 Ni, 0.055 Mg, <0.05 P and 0.37 Cu. The results are summarized in the following tabulation:

Table 5—Condensed Hardness Data, Vhn (20 Kg)

				HAZ		
Test	*Sample*	*Plate*	*Weld*	*Max*	*Avg*	*Width, in.*
1	As-cast plate, as-welded	265	227	750	429	0.093
2	Annealed plate, as-welded	173	229	665	449	0.063
3	Welded joints, annealed	165	184	335	210	0.015

The high hardness shown for the maximum in the heat-affected zone (HAZ) of tests 1 and 2 was obtained at the edge of fusion, and was confined to a very narrow zone. It reflects the presence of the carbide flash and martensite in this location. The maximum HAZ value for test condition 3 is higher than might be expected after a postweld FHT anneal. Some modification of the FHT heat treatment would be indicated for this heat of plates if further softening were desired. However, both shaper and drilling and tapping operations showed no signs of difficulty in the heat-affected zone, suggesting the FHT treatment to have produced adequate softening. The 335 Vhn value of the table for test condition No. 3 represented an extremely narrow zone, the average HAZ hardness being of the order of 210 Vhn.

The width figures shown in the table are of course approximate but reveal the somewhat narrower heat-affected zone of the as-welded joint made on annealed plate compared to that on as-cast plate, and the very narrow HAZ of the annealed joints.

The tabulation shown above represents a series of traverses on one lot of cast plates of the analysis shown. It is of course obvious that the HAZ values of the as-welded joints will vary considerably with casting composition. Elements such as manganese and nickel will be expected to increase the extent and hardness of the heat-affected zone while elements such as silicon will be expected to decrease weld hardenability. For example, a hardness survey on a welded joint made on annealed centrifugally cast Ductile Iron pipe containing 3.6 C, 2.7 Si, 0.3 Mn, 0.4 Ni and 0.05 Mg showed a maximum of 400 Vhn in the heat-affected zone, and an average value of 340 Vhn in this area.

In addition to plate casting composition, it should be recognized that any factors affecting the cooling rate such as preheat, speed of welding, heat input during welding, etc., will influence the HAZ hardness ranges.

STRUCTURE OF NICKEL-IRON ARC WELDS IN DUCTILE IRON

The weld structure contains graphite in an austenitic matrix, with some carbides co-present in cored areas. Annealing treatments result in further rejection of graphite and decomposition of any iron carbides which may be present. A carbide flash is always present at the frontier of fusion, with the adjacent heat-affected zone of the cast Ductile Iron showing considerable martensite. In multilayer welding, the heat of subsequent passes exerts an annealing effect adjacent to at least a portion of the previous weld layer. This annealing effect serves to temper the martensite and, depending upon temperature, may also decompose some of the primary carbides, yielding ferrite and small spheroids of graphite.

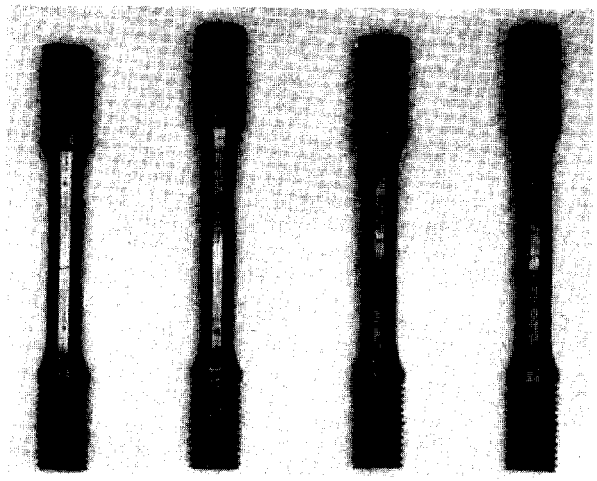
After a full anneal of the welded joint, the heat-affected zone shows mainly a ferrite matrix with spheroidal graphite. These spheroids are rather small and often quite closely spaced, interfering somewhat with obtaining of maximum ductility.

MACHINABILITY OF NICKEL-IRON ARC WELDS IN DUCTILE IRON

In general, the weld metal offers no problem in machining, except possibly under conditions of heavy plate dilution. The ease of machining of the joint is, rather, determined by the condition of the plate in the fusion zone and heat-affected zone. The structural characteristics giving rise to machining difficulties in this area have already been discussed. The carbide flash is quite narrow, usually 0.005–0.015 in. in width. The over-all width of the heat-affected zone varies, but a width of the order of 0.10 in. is common.

It has been previously mentioned that the extent and hardness of the heat-affected zone in the as-welded joint is a function of both the plate composition, heat input and cooling rate following welding. The higher the weld hardenability of the casting composition, and the faster the cooling rate, the greater will be the adverse effect on machinability, other things being the same. It is also apparent that, in multilayer welding, some amelioration of the condition in this hard zone can be effected as a result of the tempering effect of succeeding passes.

In the following illustrations, which are based on shaper machining tests on ½-in. thick ductile iron welded plate, containing about 3.60 TC, 2.40 Si, 0.30 Mn, 0.70 Ni, 0.05 Mg and <0.05 P, some of the factors influencing machinability will be demonstrated.



Weld no.	4692	4696	4693	4697
Plate cond.	As-cast	FHT	As-cast	FHT
Preheat	None	None	None	None
Postheat	None	None	FHT	FHT
Tensile strength, psi	71,500	63,500	63,500	59,750
Yield strength, psi	59,750	45,400	45,800	42,100
Elongation % in 2 in.	3	15	10	13

FHT = 1675°F.—1 hr—FC to 1275—5 hr—AC

Figure 4

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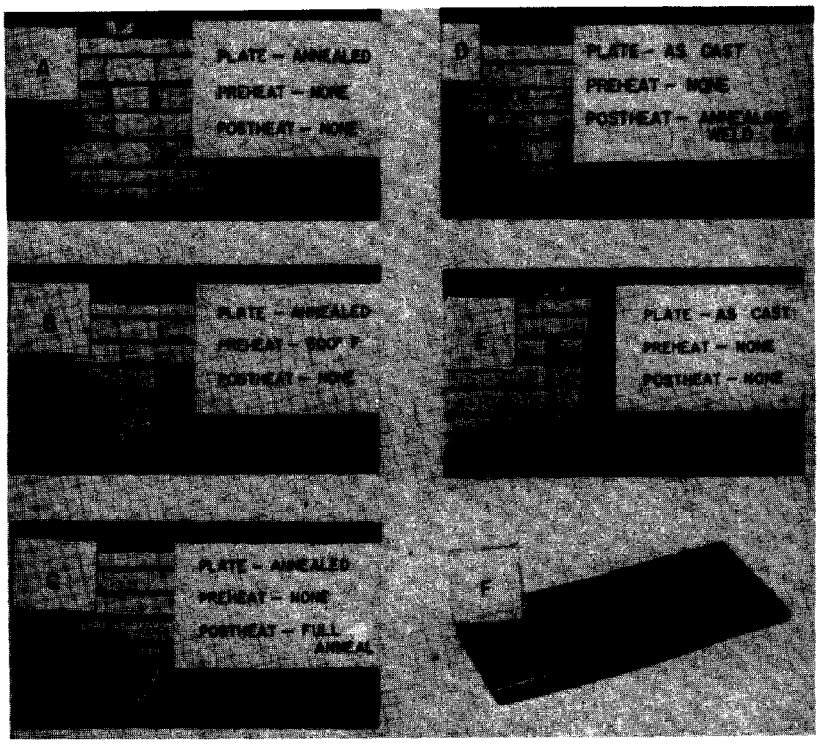


Fig. 5 Shaper machining tests on Ductile Iron arc-welded joints

The type of machining test specimen used is shown in Fig. 5*F*, the specimen, measuring 6 x 3 x $^1/_2$ in., being cut out of a 6 x 6 x $^1/_2$-in. welded joint.

The machining test was carried out in the shaper using the following conditions:

Speed, strokes per minute.....16
Depth of cut, in.....0.050
Feed, in. per stroke.....0.01
Cutting tool bit: High-speed tool steel (Rex 95)

Each cut was started at the end of the weld and continued until all but $^3/_8$ in. of the previous cut had been removed. The first cut was a reference surface cut, sufficient to remove casting skin. The balance of the cuts were 0.050 in. deep. The tool bit was resharpened before each cut except in the joints fully annealed after welding.

Figure 5*A* demonstrates the appreciably better machining response of as-welded joints on annealed plate (no cover pass). In the as-cast plate, Fig. 5*E*, tool breakdown frequently occurred even in making a single cut. The joint on annealed plate, while still showing a hard zone, could be machined without too much difficulty. By using a 600° F preheat and welding continuously, a substantial improvement can be obtained. This is particularly noticeable in welding annealed plate (*B* vs. *A*), *B* showing very little difficulty in the heat-affected zone. The effect was less noticeable in welds on as-cast plate since a pronounced hard zone persisted even with the preheat, but tool breakdown was much less of a problem. If a full anneal (1650 + 1275° F) is applied after welding, it seems to make little difference whether the weld has been made on as-cast or annealed plate, or whether or not a preheat or other special techniques have been employed. In machining the fully annealed joint, all the steps were cut without resharpening of the tool bit (Fig. 5*C*). A hole was drilled at the interface and tapped without difficulty as may be seen in Fig. 5*F*.

There is some indication that relatively low tempering treatments can produce fairly good machinability. Figure 6 shows two sections from the same joint, one portion of which has been torch annealed for 10–15 min at about 1000° F. The two cuts on each sample are 0.050 and 0.100 in. below the reference surface. The beneficial effect of the brief torch anneal is apparent. It would be expected that a 1275° F anneal (without the prior 1650° F treatment) might offer adequate machinability for some operations.

The annealing cover pass technique can also be usefully employed, but it is best applied while the joint is still hot. Figure 5*D* shows a joint made by continuous welding, a full weave annealing cover pass being applied, confining the latter to the weld proper, without remelting parent metal. Flush welds (unless a postweld heat treatment is used) are apt to be troublesome in subsequently making surface cuts. A weld reinforcement applied as a cover pass not only strengthens the joint but also tempers the hard heat-affected zone occurring as a result of depositing the flush beads (Fig. 7). The value of using a tungsten carbide tipped tool is shown in Fig. 8, the ridge effect being much less pronounced when this type of tool is used.

Summarizing the observations on machinability of welded joints in Ductile Iron, as measured by shaper tests:

1. The higher the nickel and especially the manganese content, the higher will be the weld hardenability and hence the poorer the machinability in the heat-

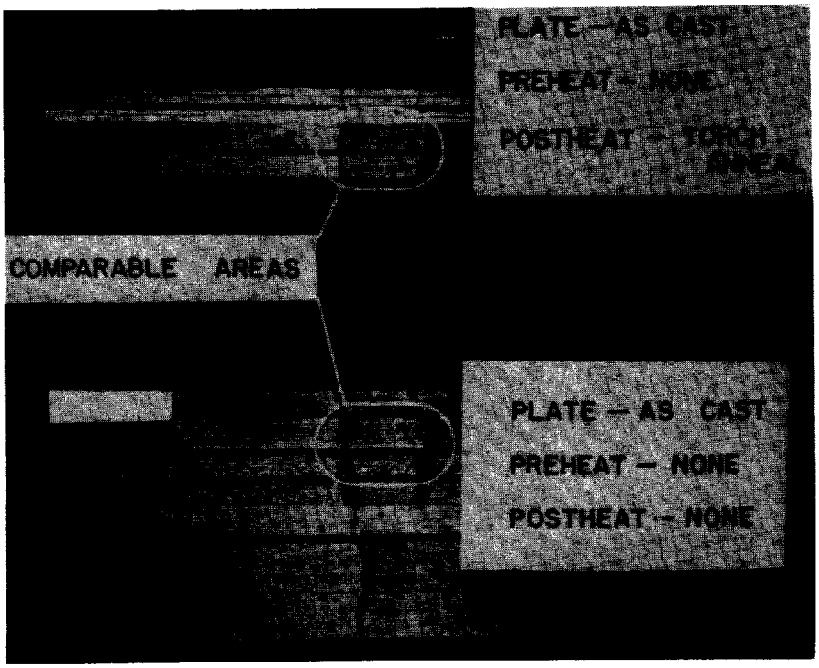


Fig. 6 Effect of a postweld torch anneal on the machinability of Ductile Iron joints

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affected zone in as-welded joints. Other alloying elements which increase weld hardenability, if present, will also be expected to decrease machinability in the heat-affected zone.

2. Joints made on plate annealed before welding are more readily machined than those on as-cast material.

3. A full anneal after welding may be expected to provide quite good machinability under wide variations of composition, prior plate condition and welding procedure.

4. A relatively low temperature postweld anneal at 1275° F will improve machinability considerably, often to a satisfactory extent.

5. A simple torch-applied postweld anneal at about 1000° F for 10–15 min ($^1/_2$-in. thick section) may provide adequate machinability for a number of operations.

6. When a measure of machinability is needed in the as-welded condition, and postweld treatments are not feasible, recourse to continuous welding, with a final cover pass, properly applied, may often suffice, when used in conjunction with a suitably chosen casting composition.

7. Carbide tipped tools offer advantages over tool steel bits, particularly where a substantial hard zone exists.

8. Where a quite high degree of machinability is a prime consideration, a full anneal should be given the welded Ductile Iron casting, prior to machining.

It is recognized that shaper operations are much less demanding than some of the other machining processes, such as lathe operations, for example, and the above observations must be considered specifically as applied to the former. It is reasonable to assume, however, that Ductile Iron joints made with nickel-iron electrodes and fully annealed after welding, should be readily machinable using most machining methods, including lathe work.

However, the data developed in the shaper tests should be used, in general, for guidance, in indicating trends in a qualitative way.

DISSIMILAR METAL JOINTS

In most types of fabrication it is often necessary for joints to be made between different materials. So too with Ductile Iron, the necessity frequently arises for it to be joined to some other material. Such a joint is readily made by metal arc welding using a nickel-iron electrode.

In Fig. 9 are shown welds wherein Ductile Iron is joined to each of five different materials: mild steel, Type 316 stainless steel, Monel, nickel and Inconel. The welded joints were 6 x 6 x $^1/_2$ in., each of these joints being made by continuous welding of annealed Ductile Iron plate to the dissimilar metal. A $^5/_{32}$-in. diam nickel-iron electrode was used at 135–140

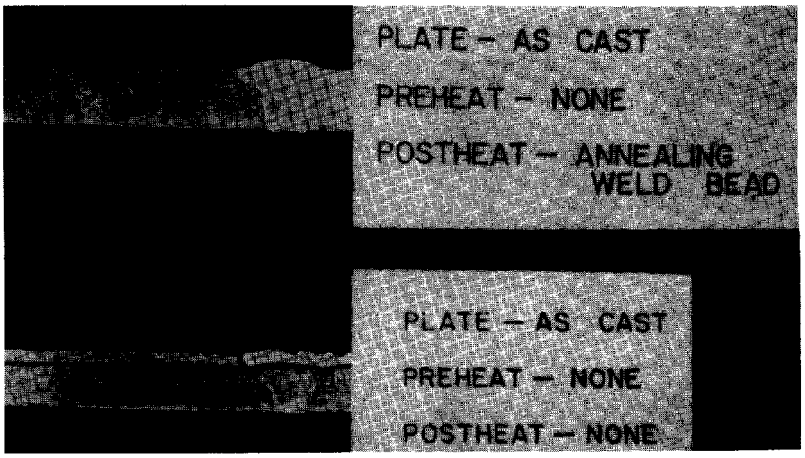


Fig. 7 Machinability of reinforced weld vs. flush weld

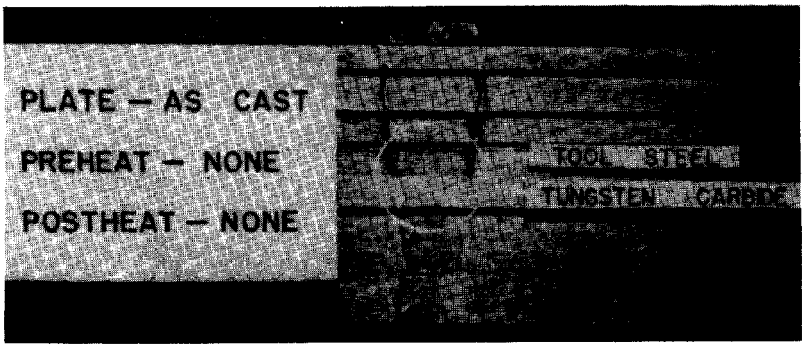


Fig. 8 Tool steel vs. tungsten carbide tools for machining Ductile Iron, as-welded

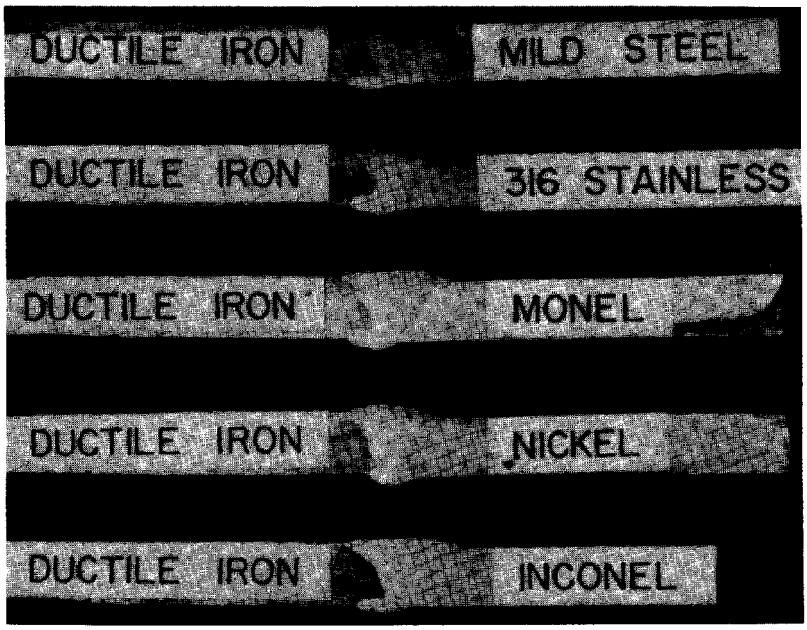


Fig. 9 Joints between $^1/_2$-in. annealed Ductile Iron plate and other metals welded with nickel-iron electrodes

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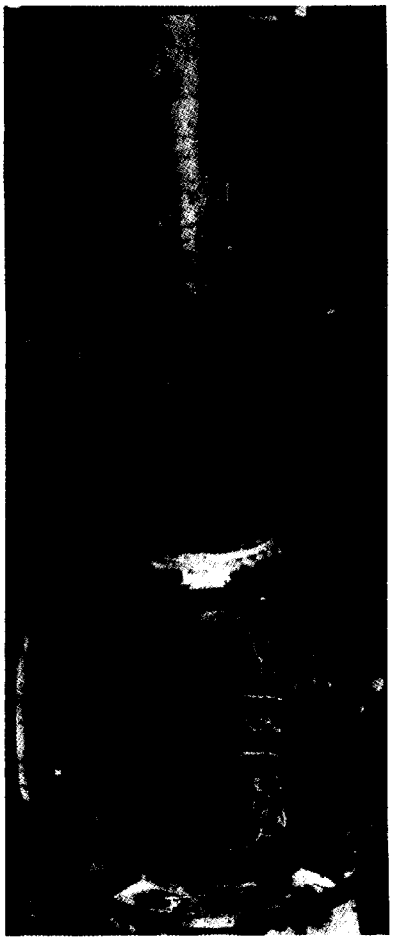


Fig. 10 Ductile Iron-to-steel weld in Hydrafuge part made with nickel-iron electrode

amp, the single-vee joint being completed with 4 passes and a sealing bead on the root side. Neither preheat nor postheat was employed for these welds. While the plates were quite severely restrained during welding, the resulting welds were free of any indications of cracking.

A great many applications involve the joining of Ductile Iron and mild steel. In Fig. 10 is shown a Ductile Iron casting welded to a steel pipe. This weldment is part of a Hydrafuge, a machine used in the pulp and paper industry. The weldability of Ductile Iron was an important factor in its selection. A nickel-iron electrode was used for this joint.

Another application involving Ductile Iron welded to mild steel is illustrated in Fig. 11. Here is shown an hydraulic farm-loader part where a cold punched Ductile Iron clevis cap is joined to a mild steel tube.

To sum up the welding of Ductile Iron to steel, it can be said that joining these two materials by welding when using a nickel-iron electrode is even more easily accomplished than welding ductile iron to itself, since the heat-affected zone of the mild steel remains soft and can yield to relieve stress, while in the matched welds, a hard brittle zone exists on both sides of the joint in the as-welded condition.

FIELD APPLICATIONS OF NICKEL-IRON ELECTRODE

Large quantities of nickel-iron electrodes sold under several trade names have been consumed in welding gray iron castings, and many case histories are available. Due to the comparative newness of Ductile Iron, the field experience with Ni-Rod 55 in the welding of this material is, of course, more limited, but a few applications can be presented. Figure 12 shows a Ductile Iron screw for a dehydrator used in the paper mill industry. The flutes of this screw are regularly overlaid using nickel-iron electrodes, in order to obtain closer clearances in the screw housing and to cut down wear. The leading face and the rim of the flutes are built up about $^1/_8$ and $^1/_4$ in., respectively, using a 200° F preheat and continuous welding. Welds are peened with a round nose tool and a pneumatic hammer.

Figure 13 shows some lengths of centrifugally cast Ductile Iron pipe welded in the annealed condition using nickel-iron electrodes. As will be seen, failure occurred in hydrostatic bursting tests at about 5000–6000 psi through the pipe wall, and a considerable distance from the weld.

Figures 10 and 11 involving field applications where Ductile Iron-to-steel joints are required have already been discussed and need no further comment.

METAL ARC WELDING WITH STEEL ELECTRODES

Field experience involving welding of Ductile Iron with steel electrodes has been scanty, and weld cracking has generally been present. The results of laboratory tests suggests that a mild steel electrode of the low-hydrogen type may be used under certain conditions. The limitations

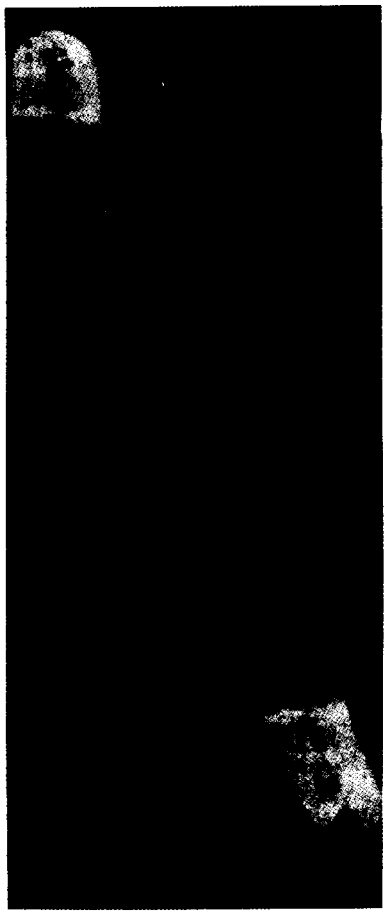


Fig. 11 Ductile Iron clevis cap, arc welded to steel tubular part

which must be recognized in order to provide a reasonable chance of success are:

1. Castings should be relatively low in manganese and nickel. The maximum permissible levels have not been established, but good results have been obtained on plates containing less than about 0.30–0.35 Mn and about 0.70 Ni.
2. Castings should be low in phospho-

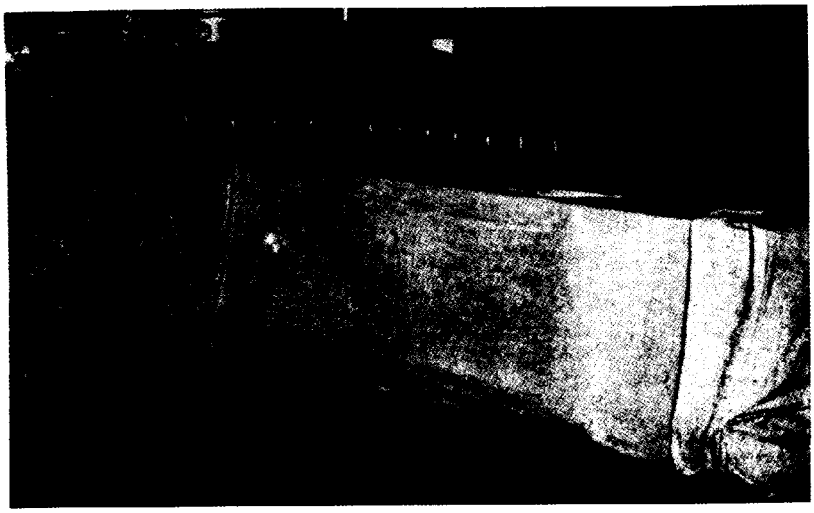


Fig. 12 Ductile Iron screw with flutes overlaid using nickel-iron electrode

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Fig. 13 Centrifugally cast Ductile Iron pipe, butt welded using nickel-iron electrode

rus, at least below about 0.05% and still lower if possible. This must be accomplished by proper selection of raw materials.

3. Castings must be in the fully annealed condition prior to welding.

The tolerance for sulfur in the base composition has not been ascertained but magnesium-treated irons are customarily low in this element and it was not a factor of concern in the present studies.

Under the above conditions of composition and heat treatment, butt-welded plates $^1/_2$ x 6 x 6 in. have been successfully produced under moderate restraint both with and without a 600° F preheat. However, cracking tendencies were observed in unpreheated welds in some instances, and the use of a 600° F preheat is advisable to provide a greater margin of safety. The commercial electrode used, though possessing quite good operating and slag characteristics on wrought steel, offers considerable difficulty with slag interference when applied to Ductile Iron. Whether the behavior is characteristic of this class of electrode or applies to the particular one used has not been definitely determined, but it is believed that considerable variations can be expected in this respect in electrodes of the low-hydrogen type obtained from various sources. Despite these operating difficulties, however, joints were successfully made on plate of the following composition: 3.68 TC, 2.34 Si, 0.23 Mn, 0.73 Ni, 0.055 Mg, <0.05 P, 0.005 S balance Fe. Tensile tests on subsize 0.357-in. diam transverse tensile test specimens showed the following range of properties: 42/47,000 psi yield strength (0.2% offset), 55/65,000 psi tensile strength and 5–9% elongation. In most instances, failure occurred near the plate-weld junction.

Attempts to weld as-cast plate of preferred composition were unsuccessful unless a high (1000° F) preheat was used and even then the physical properties were mediocre. Likewise, a Ductile Iron heat containing 1.6% nickel and 0.70% manganese exhibited weld cracking, occurring in this instance in the annealed as well as the as-cast condition. Ductile Iron containing 0.10% phosphorus, though readily welded with nickel-iron electrodes, yielded cracked welds using the steel electrode.

The field of application of steel electrodes in welding Ductile Iron is obviously much more limited than with the nickel-iron type and under the optimum conditions for their use, some weld cracking hazards are still present. Whenever steel electrodes are used, suitable exploratory tests should be made to determine if the conditions of plate composition, heat treatment and the quality of weld required will permit their use. It would be expected that steel electrodes can be more readily used to join Ductile Iron to mild steel than to itself.

Part II

OTHER WELDING PROCESSES

Ductile Iron can be welded successfully by other processes than metal arc, some of which will be discussed in the ensuing sections.

INERT-GAS-SHIELDED METAL-ARC WELDING

This generic term is cumbersome and the process is more often designated by the trade names Aircomatic or Sigma. For convenience, it will be referred to in the discussion, somewhat inaccurately but briefly, as "inert metal arc" and its counterpart with the nonconsumable electrode as "inert tungsten arc."

Inert Metal Arc: Nickel-Iron Filler Wire

This method of welding is readily accomplished using a 60/40 type nickel-iron filler wire. The wire is identical in composition with the core wire used in the nickel-iron electrode previously discussed.

Uncracked welds have been obtained readily in Ductile Iron of the preferred composition in sections up to 1 in. in thickness, without preheat or with preheats up

Table 6

Filler wire	60/40 Ni-Fe alloy
Wire diameter	0.045 in.
Wire feed	435 ipm
Welding current	170 amp, d c
Polarity	Reverse
Backing	Carbon block or copper
Shielding gas	argon
Gas flow	20 to 40 cfh
Preheat	None to 600° F (depending on job; higher preheats quite acceptable, but usually unnecessary

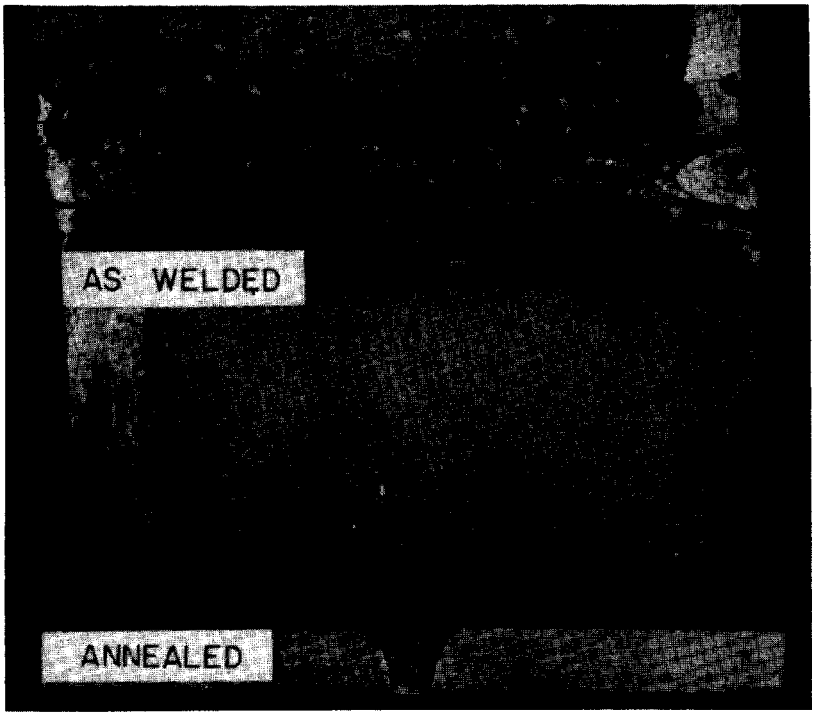


Fig. 14 Inert-gas metal-arc weld using nickel-iron filler wire

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to 1000° F, using the inert-gas metal-arc process.

In making joints in 1/2- to 1-in. thick Ductile Iron, the conditions given in Table 6 worked well.

Welds of quite good quality can be obtained with 20 cfh gas flow, but 30–40 cfh is preferred to provide greater protection against gas porosity. There is some indication that a carbon-block backing gives better protection than a copper backing.

Figure 14 shows a butt-welded joint made without preheat on as-cast Ductile Iron plate of preferred composition using the welding conditions tabulated above. A typical cross section of such a weld, after a postweld anneal (FHT) is depicted in the same figure. Good results have been obtained both with a single-vee and a double-vee or X-joint design. A number of joints were made on Ductile Iron plate containing 0.02, 0.06 and 0.10% phosphorus to examine the effect of this element, with no indications of weld or fusion zone cracking at any of the three levels.

Limited tensile test data suggest that inert metal-arc joints in Ductile Iron, made with preheats up to 600° F. using nickel-iron filler, may be somewhat inferior in mechanical properties to joints metal-arc welded with nickel-iron electrodes. For example, a joint in 1-in. plate, welded in the as-cast condition using the welding procedure of Table 6 and subsequently given a postweld full anneal yielded: 44,000 psi yield strength (.2%), 55,000 psi tensile strength and 6% elongation. The average properties of Table 4 show about 64,000 psi tensile strength and 9% elongation for this condition in metal-arc welds, with some individual tests in the latter falling to about 60,000 psi with 7% elongation.

Some tensile data were also obtained on inert metal-arc welded joints made on annealed plate of preferred composition using a 1000° F preheat and a 750° F postheat and slow cool immediately after welding. The welds were made on 1/2-in. thick annealed Ductile Iron plate. The average tensile test data obtained with the nickel-iron filler are shown below:

Table 7—Inert Metal-Arc Welds,* Nickel-Iron Filler

Post Weld, HT	*Yield strength (0.2%), psi*	*Tensile strength, psi*	*Elongation % in 2 in.*
As-welded	44,750	62,500	5
FHT	42,400	62,700	17

* 1000° F preheat, followed by slow cool from 750° F; annealed plate.

The strength properties are similar to those shown for unpreheated metal-arc welds in Table 4 and the elongations appreciably higher.

Much more test data are needed to define optimum welding procedures and to establish advantages and limitations of the process. However, the experience to date suggests that resistance to weld and fusion line cracking is quite good using nickel-iron filler wire. The process holds promise for applications where the production requirements merit its consideration.

Inert-Gas-Shielded Tungsten-Arc Welding: Nickel-Iron Filler

The response of Ductile Iron to inert-tungsten arc welding has been examined briefly employing nickel-iron filler, using the welding conditions given in Table 8.

Table 8

Tungsten electrode	3/16-in. diam
Shielding gas	Argon
Gas flow	20 cfh
Welding current	220 amp, d c
Polarity	Straight
Filler wire	3/16-in. diam
Welded plate	6 x 6 x 1/2 in.
No. of passes	2 weave, 2 lapping

Sufficient work has been done to indicate that Ductile Iron of suitable composition for good weldability will respond quite well to this process. Present information suggests that the tensile properties of tungsten arc welds made with nickel-iron filler are somewhat similar to those of the metal arc welds made with nickel-iron electrodes. Welded joints on 1/2-in. thick as-cast plate containing 3.5 TC, 2.5 Si, 0.35 Mn, 0.70 Ni and <0.05 P yielded about 67,000 psi tensile strength with 1% elongation in the as-welded condition, and about 61,000 psi tensile strength with 8% elongation after a postweld annealing heat treatment. Welded joints on annealed plate, tested in the as-welded condition, gave about 62,000 psi tensile strength with 7% elongation.

The observations on the use of the two inert arc processes are based on exploratory studies and much systematic information is obviously needed to obtain a comprehensive view of the possibilities as applied to Ductile Iron.

More needs to be known as to limitations of plate composition, conditions under which preheat should or must be used, optimum conditions of welding current, shielding gas and mechanical properties of welded joints, etc. However, the experience to date does indicate quite clearly that little difficulty should be encountered in obtaining welded joints quite resistant to weld and fusion zone cracking using both types of inert arc welding, with 60/40 nickel-iron filler wire.

It is probable that economic considerations will dictate the use of less expensive joining processes in most instances. Of the two phases of the inert arc process, the consumable electrode type is the more interesting, and may merit consideration for some types of fabrication welding, where production conditions warrant its use.

SUBMERGED ARC WELDING OF DUCTILE IRON TO STEEL

A brief study was made of this process in conjunction with an application calling for the automatic welding of Ductile Iron to steel, with the submerged arc method specified.

For laboratory studies, the test weld employed was a T-weld comprising a 12 in. long by 6 in. wide segment cut from some 3/8-in. wall Ductile Iron pipe for the base and a vertical leg of mild steel, 1/4 in. thick by about 2 1/2 to 3 in. high. This joint simulated the type of weld and the section thickness involved in the actual fabrication, though stress conditions were, of course, less severe.

A variety of nickel-iron or nickel-chromium-iron alloys were examined, most of which were inacceptable either because of excessive weld porosity or pronounced weld or fusion zone cracking tendencies, or both.

Of the various filler wires tried, the one which gave the best results with respect to weld quality was a Type 310 stainless steel wire of commercial origin containing 27.7 Cr, 21 Ni and 0.34 Si. The Ductile Iron casting contained 3.6 TC, 2.61 Si, 0.30 Mn, 0.38 Ni, 0.096 P and 0.048 Mg. Of the two proprietary melts examined, one designated "N" melt was found to be satisfactory.

The welding conditions found to give a good joint are listed in Table 9.

Table 9

Ductile Iron of low-weld hardenability, 0.10 P max
Use annealed castings
Filler wire, 25/20 stainless, 5/32 diam
Flux, N-melt (proprietary flux)
Current, 400 amp
Voltage across arc, 40 v
Polarity, straight
Speed of welding, 20 ipm
Preheat, RT to 400° F
Postheat, 850° F—1 hr—SC, applied immediately, while weldment is still warm
Arc directed more toward ductile iron member.

The immediate postheat treatment is considered important to minimize hazards of fusion line cracking. Preheat did not seem important in the size of joint made for laboratory studies, but a moderate preheat of about 400° F would probably be useful in fabrication, where welds would be subjected to higher stresses during welding.

Figure 15 shows a typical joint, with cross sections bent with face and root in tension. The fact that mild steel was

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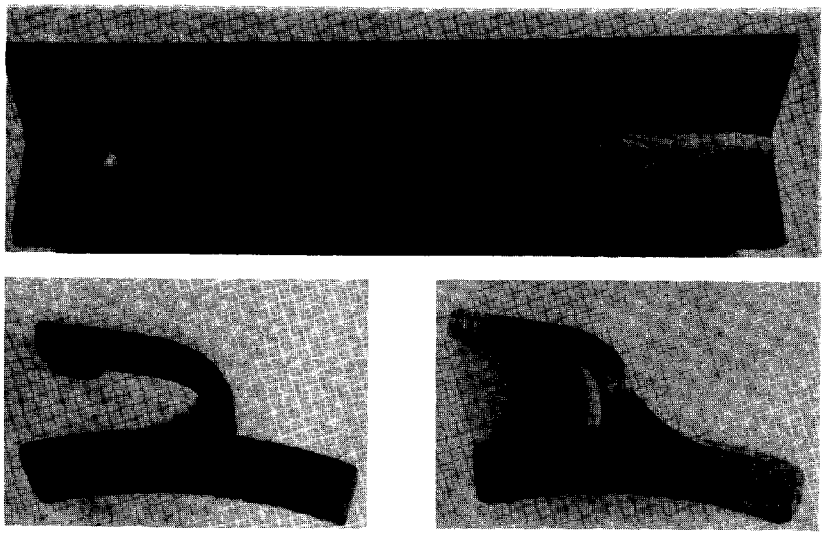


Fig. 15 Photos showing submerged arc weld of Ductile Iron to steel using 25–20 Cr-Ni filler wire

one member of the weldment undoubtedly lessened greatly the problem of avoiding fusion line cracking.

Preliminary welding tests by the fabricator interested in this job, made in accordance with the welding procedure of Table 9, were reported to be satisfactory, but, as far as is known, no commercial scale operations have yet been attempted.

While this process, as applied to Ductile Iron welding has obvious limitations, it offers some possibilities under proper conditions. Each potential application should, of course, be explored by trial welding tests, simulating the type of joint involved in the actual job.

SUMMARY

An attempt has been made to review the response of Ductile Iron to a number of arc-welding methods, including the metal arc, the inert-gas-shielded arc and the submerged-arc welding processes. Although not included in the scope of the present discussion which relates to arc welding, reference should be made to the oxy-acetylene welding process. This method, employing cast ductile iron filler rod, is being used successfully by several foundries to weld ductile iron castings.

On the basis of the present study, a few broad generalizations seem justified:

1. Ductile Iron can be welded by most of the commonly employed fusion welding processes.

2. Ductile Iron should be welded in the fully annealed condition for maximum welding response although it can be welded successfully in the as-cast condition with suitable selection of casting composition and joining process and procedure.

3. The lower the levels of those alloying elements in the casting composition which increase weld hardenability, the better is the welding response and the less important is the need for preheat.

4. Phosphorus has an adverse effect on weldability. A 0.10% maximum is permissible when welding with nickel-iron electrodes or filler wire. A 0.05% maximum should be observed if mild steel electrodes are employed.

5. The following nominal composition responded well to welding using several processes: 3.60 TC, 2.40 Si, 0.30 Mn, 0.70 Ni, <0.05 P, 0.05–0.08 Mg.

6. Metal arc welding with austenitic nickel-iron electrodes has been applied successfully to a fairly broad range of casting compositions in both the as-cast and annealed conditions, with and without preheat.

7. Inert arc welding with austenitic nickel-iron filler wire offers promise. Joints in Ductile Iron cast plate of preferred composition, welded in both the as-cast and annealed conditions were readily made in plate thicknesses up to 1 in.

8. Ductile Iron may be joined to such dissimilar metals as mild steel, stainless steel, nickel, Monel and Inconel with austenitic nickel-iron electrodes.

9. One of the above joint combinations, Ductile Iron-to-mild steel, is frequently encountered in field applications. The joint is readily made and, in general, offers fewer hazards with respect to fusion line cracking than those in which ductile iron comprises both members of the weldment.

10. Where machinability is a prime requisite, arc-welded joints should be fully annealed after welding (1650 + 1275° F treatment).

11. In the as-welded condition, joints on annealed plate are less troublesome to machine than those on as-cast plate.

12. Considerable improvement in the machining response of as-welded joints in castings of preferred composition for welding can be accomplished by special welding techniques and by brief torch annealing treatments, though machinability will be inferior to that of fully annealed joints.

13. A 1275° F postweld heat treatment (without the 1650° F prior treatment) may provide an adequate improvement in machining response for some purposes.

14. Tungsten carbide tipped tools offer substantial advantages over tool steel bits where a hard heat-affected zone exists.

15. Low hydrogen-type mild steel electrodes hold some promise for joining Ductile Iron, provided a low-alloy, low-phosphorus casting composition is used and the castings are welded in the fully annealed condition.

16. Submerged arc studies were confined to joining of Ductile Iron to mild steel. The process may be useful for special applications but its field appears limited on the basis of present knowledge.

A great deal of systematic study is needed in all the above fields to evaluate the advantages and limitations of the various arc-welding processes, to establish preferred casting compositions for maximum welding response and to develop improved welding procedures.

However, it is hoped that these observations and experiences will serve to demonstrate that Ductile Iron can be welded quite readily by several methods and will encourage further studies on the welding of this new engineering material.

ACKNOWLEDGMENTS

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WELDING DUCTILE IRON CASTINGS

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Approximately 8 million tons of ductile iron castings are produced on a worldwide basis and, in achieving this position, ductile iron has taken markets from steel and malleable iron castings as well as steel fabrications. The reasons for these conversions are several but in most cases it is found that ductile iron castings meet the property and service requirements demanded at lower cost to the user.

Whereas ductile iron possesses many attractive decisive features, one area where it is at a disadvantage in comparison to steel and certain grades of malleable iron is the greater difficulty involved in welding ductile iron castings. Nevertheless, by using suitable welding processes and techniques it is quite possible to produce ductile iron weldments with good properties which are adequate for many purposes.

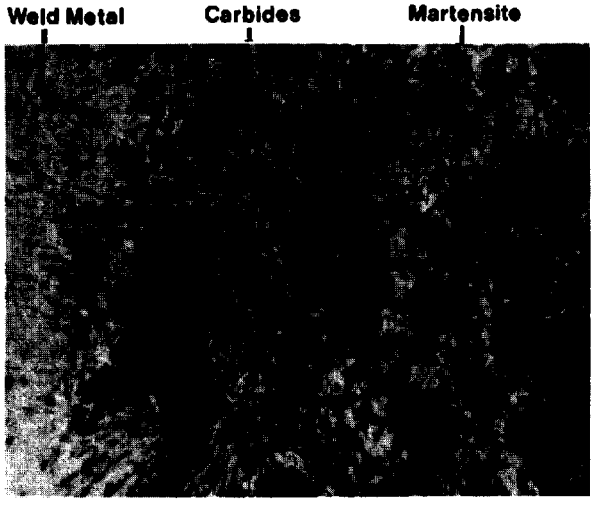


Fig. 1. *Typical microstructure of metal-arc welded ductile iron in the area of the weld metal, weld metal/nodular base junction and heat-affected zones.*

Indeed there is an urgent need for greater attention to be paid to the available joining techniques for castings since, although the casting process is extremely flexible and a single intricate ductile iron casting is frequently preferable to a fabrication or assembly, there are situations where the joining of ductile iron castings to themselves, or to components in other materials, is advantageous. Designers, value engineers and production engineers are becoming increasingly interested in situations where:

a) to meet certain service requirements a composite component of ductile iron and steel is required—particularly the ability to join ductile iron to conventional wrought forms such as tubes, bars and angle sections.
b) the design of the component is so intricate or large that a one-piece casting is either impractical or uneconomic.

Additionally there are several situations in the foundry where the repair of certain casting defects by welding can recover an otherwise scrap casting.

For all of these reasons therefore it is necessary for foundrymen, designers and users of ductile iron castings to become familiar with the problems, processes, and techniques involved in the welding of ductile iron.

Metallurgical Considerations

Fusion welding involves melting of the filler (weld) metal as well as an area of the base metal. Low carbon steels and surface de-carburized grades of malleable iron do not suffer the problems arising from the formation of hard, brittle constituents on re-solidification. It is the high carbon content of grey and ductile cast irons which is the principal cause of their weld-

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ing difficulties.[1] As a result of the high carbon content, eutectic carbide (cementite) as well as martensite and bainite can be found in the weld metal/base metal junction and heat-affected zone regions (Fig. 1). The formation of hard and brittle microconstituents in ductile irons can cause serious difficulties in machining and can result in internal stressing, distortion and cracking.

The problems outlined above are of metallurgical origin and for satisfactory welds to be achieved, attention in the following areas is required:

1. Grade and composition of the base ductile iron.
2. Material selection of the welding rod/wire.
3. Control of the thermal history—before, during and after welding.

Grade and composition of the base ductile iron

Ferritic matrix ductile irons are generally regarded as the most suitable of the range from the viewpoint of weldability. Clearly, a ferrite matrix is the closest approximation to a mild steel structure and will involve less problems due to hardening in the heat-affected zone than will a pearlitic matrix ductile iron. This is due to the fact that the element which has the greatest influence on the temperature at which martensite formation commences (Ms) is carbon. The Nehrenberg formulae for calculating the Ms temperature in ductile iron is given as follows:

$$Ms\ (°C) = 500°C - 300\ (\%\ C) - 33\ (\%\ Mn) - 22\ (\%\ Cr) - 17\ (\%\ Ni) - 11\ (\%\ Si) - 11\ (\%\ Mo)$$

where the elements are percentages present in the matrix.

The carbon content of the ferrite matrix is extremely low and, with good quality/high purity charge materials, the levels of chromium and manganese can also be maintained very low.

The Ms temperature of ferritic ductile iron is higher than in pearlitic nodular. This is advantageous since it is known that the volume change associated with the martensite transformation is greater with lower Ms temperatures. Also the hardness of martensite is directly proportional to the carbon content—again favoring the ferritic grades.

The influence of silicon content is also of importance in considering the base ductile iron. Silicon is a strong graphitizing element, tending to raise the austenite/graphite eutectic temperature and lower the austenite/carbide eutectic temperature. Consequently, from the viewpoint of avoiding carbide formation (or rendering any carbides formed less stable at elevated temperature) the silicon content of the ductile iron should be as high as the specification and application will allow.

While it is evident that ferritic ductile iron presents less metallurgical difficulty in welding, it must be stated that satisfactory welding of pearlitic grades can and is being achieved as will be shown later.

The diffusion of carbon in austenite at the temperatures involved in the zones adjacent to the weld is high. Dissolution and subsequent diffusion of carbon from the graphite nodules into the austenite continuously occurs in the welding of all grades of ductile iron.

With regard to the graphite nodules present in the structure, there is growing evidence[2] to support the statement that a high nodule count in the base ductile iron is beneficial and reduces the danger of carbide formation on re-solidification.

Selection of filler material

Mild steel filler rods/wires are readily available at relatively low cost. In general however, steel filler materials have not been successful with ductile iron due to the rapid absorption of carbon resulting in hardening and cracking. Recently however, Schock[3] has reported successful welding of ferritic ductile iron using steel electrodes.

The filler materials which have been mainly employed therefore are those which are capable of absorbing the carbon and retaining it in a form which will not be detrimental to the properties of the weld. For this reason, nickel and nickel/iron electrodes are most widely used. Nickel carbide (Ni_3C) is much less stable than iron carbide (Fe_3C) and under most solidification conditions the carbon separates in the form of graphite.

Control of thermal conditions

It is clear from the foregoing discussion that if reasonably consistent properties are to be obtained from a weldment involving ductile iron, then some degree of control over the heat input and cooling rate is essential.

Preheating

Preheating is normally advisable for the following reasons:

a) to prevent cold cracking.
b) to reduce hardness in the heat-affected zone.
c) to reduce residual stresses.
d) to reduce distortion.

Either the entire part to be welded may be preheated or, if local preheat is carried out, only the vicinity of the joint is heated. Almost any type of furnace can be used for preheating provided that the prepared areas of the welds are not contaminated by soot or smoke during the heating cycle. If furnaces are not readily available, castings can often be preheated by using town gas or propane burners or an oxy-acetylene torch. When these methods are employed, it is advisable to surround the castings with firebricks or asbestos wool or board which will assist temperature equalization throughout the casting.

When local preheating techniques are employed, extreme care is required in the initial stages of applying the heat to the casting in order to avoid cracking due to thermal shock. This is particularly valid for complex contour pearlitic ductile iron castings.

Correct temperature control is important in preheating. This is relatively easy to achieve where good control over heating furnaces applies but if the castings are preheated by gas burners or torches then surface contact pyrometers or heat sensitive crayons/paints are suitable.

Pearlitic ductile iron should be carefully preheated to about 320–340°C. Ferritic grades normally require only mild preheating in the range 100–200°C.

In some cases, in order to eliminate problems of carbides, a deposit of weld metal is put down on the prepared weld faces and the piece is then fully annealed or normalized to decompose any carbides. The "coated face" castings are then ready for welding without involving any further fusion of the base ductile iron.[1]

Minimizing heat input during welding

This can be approached in different ways depending on process:

a) non-fusion methods—welding or joining can be achieved by non-fusion techniques, such as "Powder Welding", such that the eutectic temperature is never exceeded and consequently eutectic carbide does not form.
b) limiting the time available for the austenite to take graphite (carbon) into solution. This involves welding by a process which gives a high ratio of deposition to heat input.

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Controlled post-weld cooling rate

The rate of cooling immediately after metal-arc welding can be in the order of 300°C per second. In order to achieve thermal decomposition of carbides which have been formed and avoid the formation of martensite, efforts must be made to slow down this rate of cooling or even to transfer the weldment to an annealing furnace if this is possible.

Practical Aspects Involved

Weld preparation is a fundamental prerequisite to the production of a sound weld. This aspect cannot be emphasized too strongly and one of the most frequent causes of an unsatisfactory weld is lack of care and attention to the preparation of the faces to be joined.

Surface preparation

1. The casting skin must be removed in the vicinity of the intended weld junction.
2. The joint faces must be freshly ground or machined.
3. Any rust, dirt, grease, oil or scale must be removed from the joint faces.
4. In the case of a defect salvaging repair, the defect must be removed completely and a clean metal surface presented for welding.

These points regarding preparation should be rechecked immediately prior to the welding operation.

Design of joint

The design of the joint is dependent upon factors such as metal thickness, casting geometry, welding process, service application (whether fabrication or repair), etc.

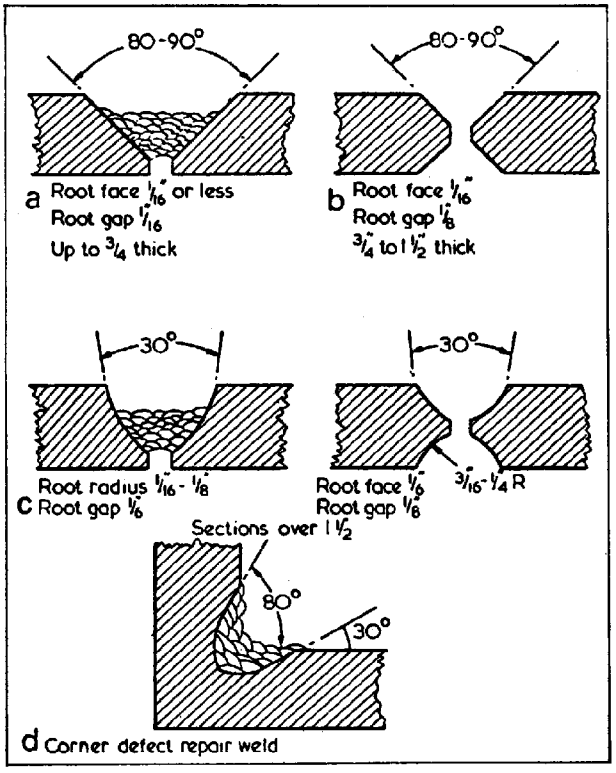


Fig. 2. *Examples of joint preparations and designs.*

Fig. 2 illustrates typical joint designs which can be recommended.

In addition to design of the joint itself, it is strongly recommended that if a welding operation involving a casting is envisaged at the original design stage, then the designer should give consideration to the various factors involved.

Casting section: should generally be not less than ¼ in. (6 mm)

Locating: The two components should locate together as accurately as possible. Machining is preferred for this purpose. The main "joint preparation" can be designed into the casting.

Accessibility: The welder must be allowed a clear access to the weld and most welds will be conducted from above on to a horizontal weld.

Loading on weld: The designer should take every opportunity to minimize the service stresses on the weld.

Distortion: By careful programming of the welding procedure, distortion can be minimized.

Selection of welding process

The selection of the welding process, ideally, should take several factors into consideration:

a) Service application of the weldment.
b) Equipment available and whether the welding is to be performed in the factory or on site.
c) Number of welds to be made and speed of operation required.
d) Whether positional or overhead welding is required.
e) Whether machinability of the weld is required.
f) Extent of welding defects which can be tolerated and quality control standards involved.

There is a variety of welding processes available which broadly fall into two groups:

a) **Electric welding:** The two most important electric welding methods are 1) manual metal-arc welding using conventional flux covered consumable filler metal electrodes, and 2) automatic and semi-automatic gas shielded welding using thin wire consumable filler materials (MIG welding).
b) **Gas welding:** The two most important gas welding processes are 1) oxy-acetylene gas torch fusion welding using filler rods and welding fluxes, 2) powder welding—non-fusion, oxygen gas torch, system using fine powdered alloy as the filler.

These processes will now be discussed in detail.

Electric welding

Metal-arc welding

Metal-arc welding is probably the most common welding method employed. This is particularly true of casting reclamation welding in foundries, but also a considerable amount of mild steel components are joined to ductile iron castings under controlled conditions of metal-arc welding.

The process is basically simple—an electric arc is struck between the consumable filler electrode and the workpiece, the resulting heat generation being sufficient to cause fusion. Because of its basic simplicity and its wide use in steel fabrication, the approach of welders to metal-arc welding of cast iron materials is occasionally lacking in care and detailed consideration.

Austenitic grades of ductile cast iron are susceptible to cracking in the heat-affected zone when metal-arc welded and

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Table 1

SOME MECHANICAL PROPERTIES OF BUTT-WELDS MADE IN PEARLITIC AND FERRITIC S.G. IRONS USING NICKEL-IRON FLUX-COATED ELECTRODES[4]

Grade of iron	Condition of cast plate	Mechanical properties							
		Yield strength, tons/in.²	Maximum tensile strength, tons/in.²	Elongation, %	Unnotched Charpy impact strength, ft. lb.*	Hardness, HV		Bend angle, ° †	Nil-ductility transition temperature from drop-weight specimens ‡
						Weld metal	Maximum in heat-affected zone		
Ferritic	Unwelded	15–20	24–30	12–20	100–106	—	—	78–80	– 120 °F
	As-welded	18–26	23–31	1–9	14–20	205–250	650–700	14–28	—
	Stress relieved at 600–650 °C for various times	16–26	24–32	5–14	28–36	185–235	400–430	16–38	—
	Full ferritise anneal — 2–4 h 900 °C 8–16 h 700 °C furnace cool	16–23	27–30	6–12	30–36	175–185	160–180	40–45	– 95 °F
Pearlitic	Unwelded	30–39	37–47	3–5	10–12	—	—	5–10	—
	As-welded	30–34	30–37	1–3	12–15	190–225	650–700	14–16	—
	Stress relieved at 600–650 °C for various times	30–32	28–33	1–3	10–12	175–205	420–500	—	—
	Full ferritise anneal — 2–4 h 900 °C 8–16 h 700 °C furnace cool	22–24	26–29	6–12	38–40	175–180	180–190	38–42	—

* Tests made at room temperature.

† The bend angle quoted covers specimens of different shapes and with both weld face and weld root used as the point of application of load.

‡ In this test a specimen 14 in. long is supported on rests and a weight of standard size is dropped from a specified height. A notch is provided, usually by placing a brittle weld bead across the center of the specimen, which introduces a running crack after about 3° bend. A stop prevents bending beyond 5°. Thus, if the specimen breaks during the test the mechanism of failure is primarily cleavage. Complete failure of drop-weight specimens occurs at temperatures below the "nil-ductility transition temperature".

because of this the process is not generally recommended for joining austenitic castings to one another or to austenitic steel.[4] This tendency to cracking in austenitic irons is reported[5] to be greatly influenced by the level of magnesium, silicon and phosphorus in Ni-Resist D2 but satisfactory welds are achieved when these elements are held below 0.06 percent, 1.5 percent and 0.03 percent respectively.

Metal-arc electrodes in 55 percent nickel-iron alloy are generally preferred when welding nodular cast iron. Nickel-iron electrodes with specially formulated electrode coatings designed to give a soft, non-penetrating arc are commercially available and the low heat input associated with the use of these electrodes is claimed to minimize the formation of hard and brittle structures in the heat-affected zone.[6]

For austenitic ductile iron castings a 20 percent nickel-iron-magnesium electrode, designed to have solidification charac-

Table 2

TENSILE PROPERTIES OF BUTT-WELDS MADE BETWEEN S.G. IRONS AND MILD STEEL BY THE METALLIC-ARC PROCESS USING FLUX-COATED NICKEL-IRON ELECTRODES[4]

Initial structure of S.G. iron plates	Condition of weldment	Mechanical properties				
		Yield strength, tons/in.²	Tensile strength, tons/in.²	Elongation, %		Position of fracture
				½ in.	2 in.	
Pearlitic	1 in. thick plates, as-welded	18.7	28.2	6	4.5	Weld metal
		22.0	31.0	52	17	Mild steel
		17.1	30.6	58	17.5	Mild steel
Ferritic	1 in. thick plates, as-welded	17.0	27.1	6	5	Weld metal/S.G. iron
		18.5	30.2	4	10	Weld metal/S.G. iron
		18.5	30.3	—	13.5	S.G. iron
		19.0	29.8	6	9.5	Weld metal

Additional welding data: Double V-joint, included angle 75°.
Pearlitic S.G. iron plates preheated to 200 °C before welding.
Root runs, 10 gauge electrode 65 A, a.c.
Subsequent runs, 8 gauge electrode, 120 A, a.c.
Tensile test pieces, 0.505 in. dia., machined across the joint.

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Fig. 3. *Stub axle (ferritic ductile iron) welded by metal-arc process to mild steel tumbling barrel.[6] (See text for description.)*

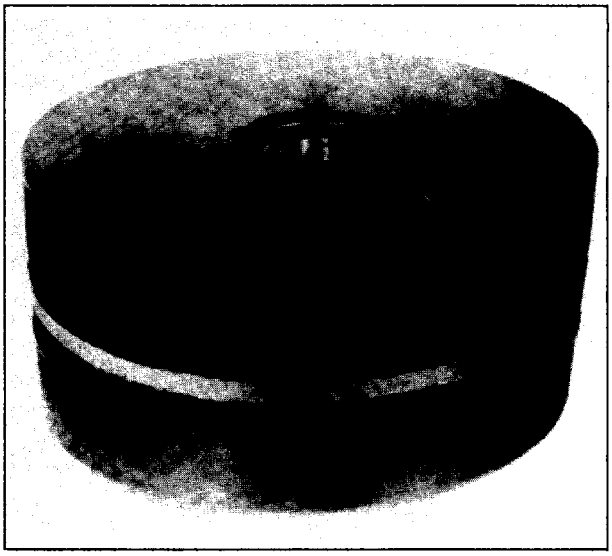


Fig. 5. *2-piece (pearlitic ductile iron castings) piston welded by metal-arc process.[6] (See text for description.)*

teristics exactly matching those of the basic material, is available which is claimed[5] to significantly reduce the dangers of cracking.

Mechanical properties and examples

The major disadvantage of metal-arc welding for ductile cast irons is the fact that it is a fusion process involving high localized heat input. Despite claims made for certain non-ferrous electrode compositions, the process is usually characterized by a well-defined band of carbide at the base metal/weld metal interface. These carbide regions are invariably associated with regions of martensite and bainite. The extent of these carbides and hard regions is minimized by using as low a power input as possible compatible with the maintenance of adequate metal fusion and depositing the filler metal in short runs.

The tensile strength and elongation of butt-welds made in Meehanite nodular iron using Ni/Fe electrodes were reported by Forrest[1] to be approximately 60% of base metal strength for the pearlitic nodular grade, Type SP (increasing to 66% after heat treatment at 910°C) and 82% of base metal strength for the ferritic nodular grade, Type SF (increasing to 86% after heat treatment)—ductility reached only 3% elongation in both cases. These figures correspond closely with results reported[7] by the American Welding Society where, for a ferritic ductile iron, welded with Ni/Fe rods, 76% of average base metal strength was achieved with severe reductions in elongation value. Typical mechanical property figures for SG iron

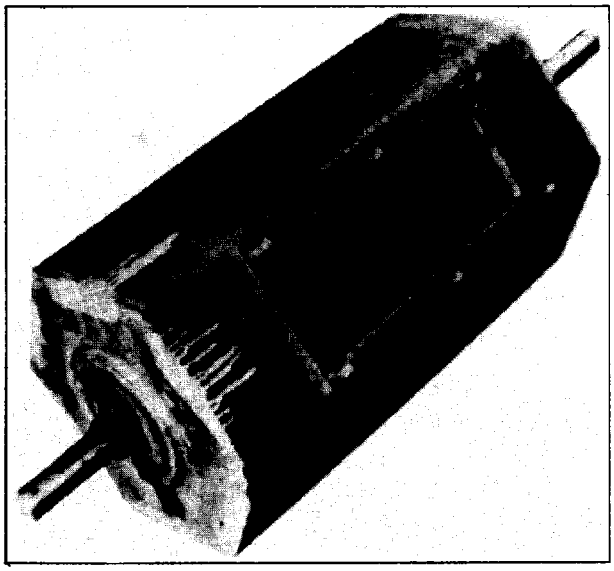


Fig. 4. *General view of tumbling barrel with welded stub axle assembly (as Fig. 3 above).*

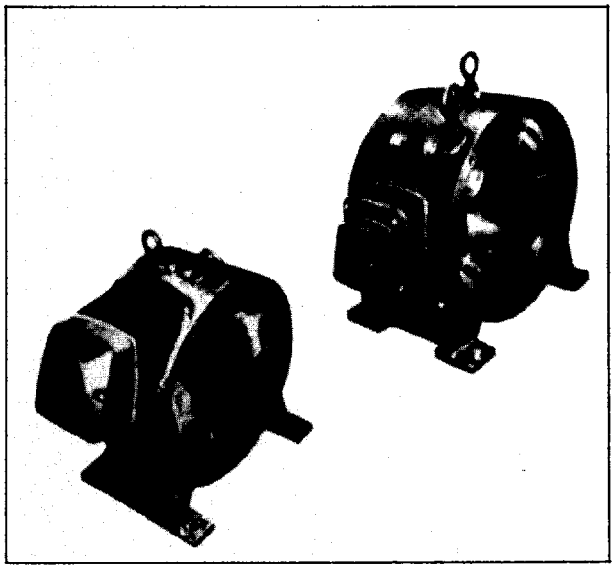


Fig. 6. *Ductile iron terminal box castings welded to steel frames by metal-arc process.[6]*

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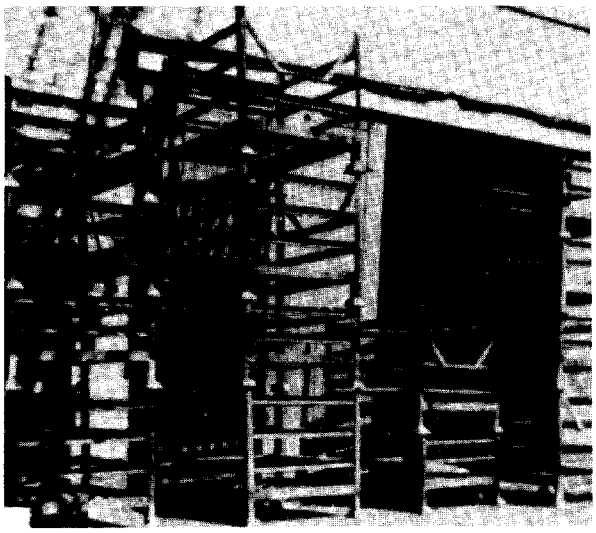


Fig. 7. *Storage stillages[8] with metal-arc welded ductile iron feet.*

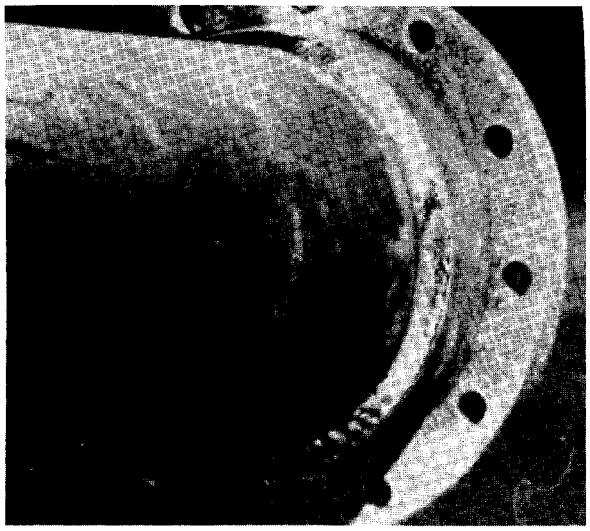


Fig. 9. *Ductile iron flange welded to centrifugally cast ductile iron pipe.[8] (See text for description.)*

metal-arc welds were assembled by Morgan[4] and are given in Tables 1 & 2.

Despite the fact that tensile strengths approaching base metal strength can be obtained, the presence of carbides and hard transformation products is undesirable in components subject to impact loading or high fluctuating stresses. Examples of metal-arc welds of ductile iron are shown in Figs. 3 to 12.

The heat-treated ferritic ductile iron stub axles and ductile iron sub-assembly castings, 6 and 20 mm thick, shown in Fig 3 & 4, are welded to each other and to mild steel by the metal-arc process, to make components for overhead traveling crane lifting hoist carriage assemblies and the components are subject to dynamic and static stresses. The metal-arc welding is carried out using 3 and 4 mm diameter pure nickel and Ni/Fe electrodes with no preheating or heat treatment after welding. Several hundred of these assemblies have been made for commercial use.

The horizontal air-compressor piston shown in Fig. 5 is a metal-arc welded assembly of two ductile iron pistons. The mating surface and weld bevels of the piston halves are machined, bolted together on an arbor and furnace preheated to 565 °C. The preheated assembly is then welded by metal-arc using Ni/Fe electrodes and subsequently heat-treated to a pearlitic matrix structure to meet the tensile requirement of 560N/mm² tensile, 420N/mm² proof strength, 3% elongation and 201–269 Brinell hardness. By casting the pistons in two halves and later joining them by welding, several production problems were minimized. The light-weight piston in ductile iron with 6 mm wall thickness replaced a heavier grey iron piston having 16 mm wall thickness to reduce inertia as speeds

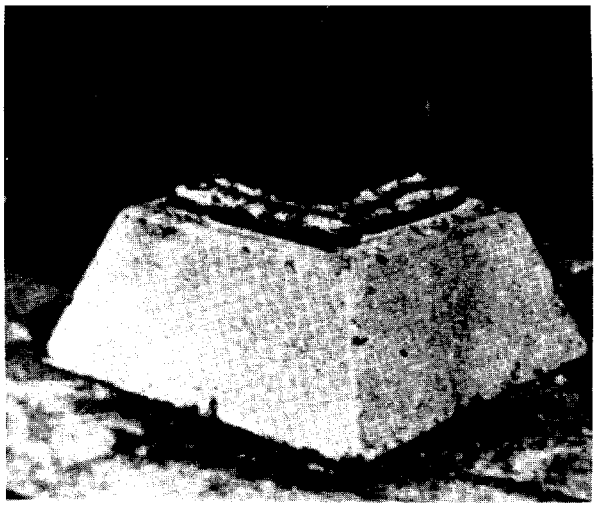


Fig. 8. *Close-up of welded ductile iron foot.*

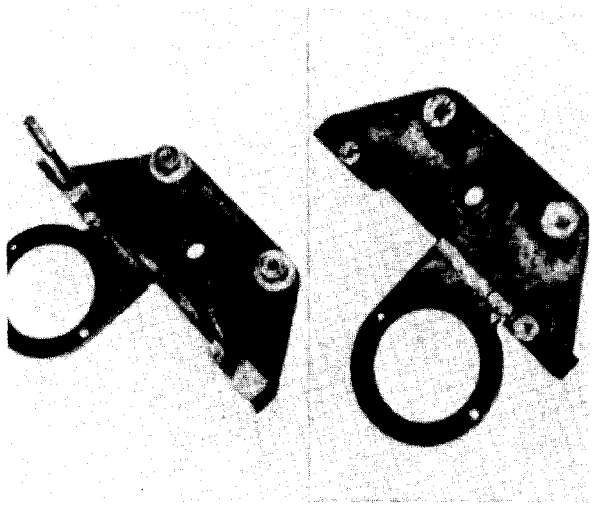


Fig. 10. *Ductile iron stub axle assemblies, metal-arc welded to steel for overhead lifting hoist.[8]*

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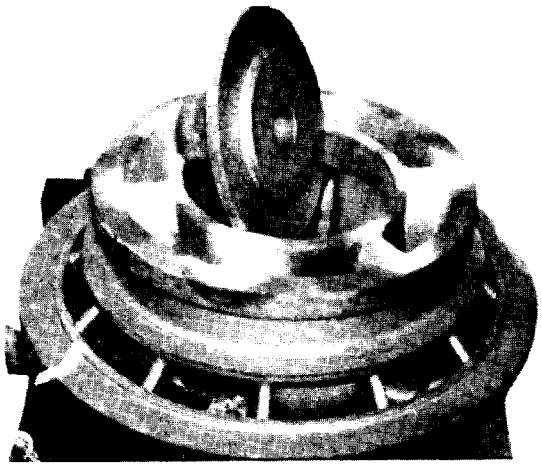


Fig. 11. *Ferritic ductile iron holder and duct castings for a supercharger welded by metal-arc process. (See text for description.)*
Courtesy: Cooper Bessener Co., Ohio, USA.

were increased.

Fig. 6 shows ductile iron terminal box castings welded to steel frames by metal-arc process using 55 Ni/Fe electrodes.

For added strength, ductile iron feet are welded to a range of fabricated steel stillages as shown in Figs. 7 and 8. The welding is carried out by metal-arc welding using Ni/Fe electrodes. Hundreds of such fabrications have been made for commercial use.

The centrifugally cast ductile iron pipe shown in Fig. 9 has a sand-cast ferritic ductile iron flange welded on by the metal-arc process to produce a pipeline junction. Pipes of varying sizes from 80–400 mm dia. were welded in this manner using nickel-coated electrodes. Up to 300 mm dia. no preheating was applied but above 300 mm dia. the flange was preheated to about 450 °C. No post weld heat treatment was applied and all welded pipes were tested to 20 atmospheres pressure.

Another good example of a metal-arc weld of ferritic ductile iron is shown in Fig. 10. Here, ferritic ductile iron stub axles and the sub-assembly castings (also ferritic ductile) welded together and also to mild steel to produce components for overhead traveling assemblies. The components were subjected to both static and dynamic stresses. No preheating or post-weld heat treatment was performed and the metal-arc welding was carried out manually using 3 and 4 mm dia. pure nickel and Ni/Fe electrodes. Several hundred of these assemblies were produced satisfactorily.

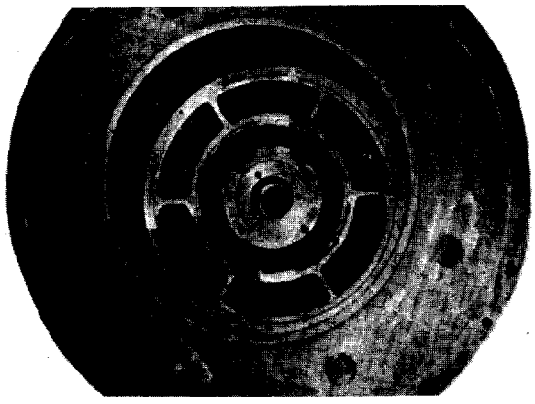


Fig. 12. *Same as Fig. 11.*

The final example of metal-arc welded ductile is the gas duct of a supercharger shown in Figs. 11 and 12. The bearing holder is welded into the duct casting, both of which are ferritic ductile iron. The photographs show the assembly before and after welding and machining (the bright spots on the top surface are dye marks for testing soundness of casting). The metal-arc welding was made with 55 Ni/Fe and nickel electrodes after preheating the castings to 260 °C. In service the assembly must withstand temperatures of 600 °C.

Recommended process techniques

Equipment required: Standard welding transformer. Electrodes—usually Ni or Ni/Fe and either flux coated or with fluxed core.

Power settings: In accordance with the electrode manufacturer's instructions, but generally as follows:

Electrode Size (dia.)	D.C. (reversed polarity)	A.C.
	Amp	Amp
1/8 in. (3 mm)	80–100	90–110
5/32 in. (4 mm)	100–120	120–140
3/16 in. (5 mm)	120–150	130–160

Preheat locally, or for small castings totally, to about 250 °C. The lowest possible amperage setting should be employed, compatible with maintaining a satisfactory weld deposit, to minimize the heat input.

The first run (root run) is most important and must be entirely sound and free from defect, slag, etc. Wherever possible, it is advised that back grinding of this deposit should be performed to remove any defective regions before proceeding with the remainder of the weld. Slag must be peened and brushed away entirely from each deposit.

For the majority of commercially available rods, short straight deposit beads are recommended. Each run should be allowed to cool somewhat before proceeding. The arc should always be struck at a previously deposited bead and not at the base iron.

When the weld is completed, the weldment should be allowed to cool slowly by means of insulation. Preferably, if dimensions permit, the weldment should be transferred to an oven or furnace at about 300 °C and allowed to cool in the furnace. With the ductile grades, particularly the ferritic ductile grades where ductility is required, it is advised to transfer immediately after welding to an annealing furnace to remove the carbides formed during welding.

Overlay or "buttering" techniques are frequently employed. This involves depositing a weld layer onto the joint faces prior to actually making the joint. These layered faces can be annealed to remove the carbides and then the joint weld made. This technique has the advantage that welds can be made "on site" where annealing facilities are not available.

Short-arc MIG welding

Short-arc MIG welding, or to give it its full description, consumable metal electrode, inert gas-shielded welding, is a versatile process which has developed over a period of several years for application to a variety of steels and non-ferrous alloys. As a result of the need to minimize heat input, in order to avoid the undesirable metallurgical consequences, the process was

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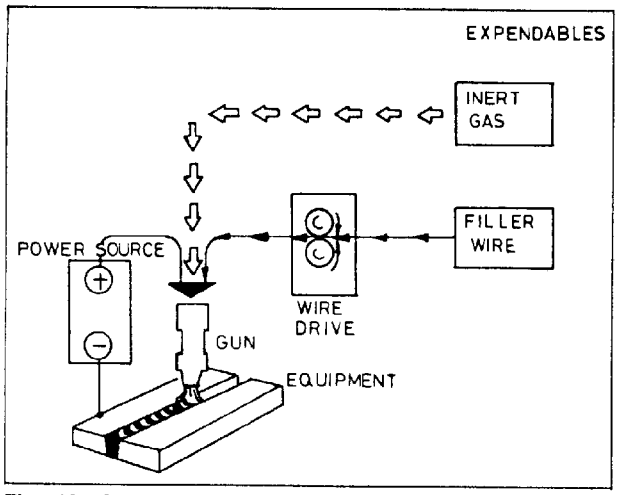


Fig. 13. *Schematic illustration of short-arc MIG welding arrangement.*[4]

modified to become a low heat input process—short-arc MIG.

In short-arc MIG welding, the arc is struck between a relatively fine gauge filler wire (about 0.03 in. diameter) and the workpiece (Fig. 13). The wire is fed continuously on to the workpiece where it acts as a fuse, melts, and the filler metal droplet becomes detached. This is automatically repeated several times each second and, consequently, the mode of metal transfer is termed "dip transfer". Throughout the process, the wire and the weld pool are shielded with an inert gas (Argon).

Low currents and voltages are used, e.g., 160 A at 20 V, and the power source is of a constant voltage type which ensures that the metal transfer from the filler wire is brought about by melting during rapid short circuits between the wire and the weld pool.

Because of the high ratio of deposition rate to heat input, the weld pool is small in comparison to that produced by conventional consumable metal-arc processes. Thus welds can be achieved in all positions and welding stresses and distortion are reduced. Also, since there is no need for interpass slag removal, the short-arc process is a fast welding method suited to production lines.

In comparison with normal metal-arc welding, the high ratio of weld metal deposition rate to heat input with short-arc MIG welding results in only discontinuous carbide distribution and narrower regions of the heat-affected zone. Since there is relatively small base metal dilution with the nickel alloy (Ni 61) filler wires, the carbides which do form tend to be restricted to regions surrounding the graphite nodules.

Table 3
MECHANICAL PROPERTIES OF SHORT-ARC MIG WELDED MEEHANITE METAL

Plate 1 in. < 6 in. x 6 in.	Filler Metal	Ultimate Tensile ton/in.	Elongation %
Pearlitic Nod. Meehanite Type SP unwelded (normalized)	—	56.3	2.4
Type SP welded to Type SP	Nickel 61	26.4 29.5	1.2 1.5
Ferritic Nod. Meehanite Type SFF unwelded	—	29.3	22.5
Type SFF welded to Type SFF	Nickel 61	29.2	5.5
" " "	" 61	29.0	5.9

Mechanical properties and examples

Previous assessments[1] of short-arc MIG welding of Meehanite nodular irons Types SP and SFF are shown in Table 3 and it can be seen that the fully ferritic ductile plates achieved 100% of the base metal strength but elongation was reduced from 22.5% to just less than 6%.

Somewhat better results were reported by Morgan[4] as shown in Table 4 where elongation figures of 11½% in welded ferritic iron plates were obtained.

Fig. 14 illustrates a short-arc MIG welded piston for a compressor. The two machined halves (ferritic ductile iron castings) were heat-treated prior to welding to conform to the specification calling for 40 N/mm² tensile with 15% elongation but no preheating was employed prior to short-arc MIG welding using 0.8 mm dia. wire (Ni 61) under argon gas shielding. Approximately 5000 of these pistons were reported to be in commercial use.

Manganese steel wear plates were welded on to ferritic/pearlitic ductile iron railway axle boxes by short-arc MIG welding as shown in Fig. 15. No preheating or post-weld heat treatment was applied. Short-arc MIG welding using 0.8 mm dia. pure nickel wire under argon gas protection was used. Approximately 25,000 axle boxes with welded-on plates are known to be in service.

Short-arc welding was again used to attach the ferritic ductile iron coupling to the steel operating arm of a hydraulic jack

Table 4
TRANSVERSE TENSILE PROPERTIES OF BUTT-WELDS IN 1 IN. THICK PEARLITIC AND FERRITIC S.G. IRON PLATES—"SHORT ARC" PROCESS[4]

Plate	Filler metal	Transverse tensile properties* 0.2% Proof stress, tons/in.²	Maximum stress, tons/in.²	Elongation over weld, % 2 in.	Location of fracture
Pearlitic as-cast	Unwelded	25-30	40-50	1-3	—
	Nickel 61	23.2 23.2	33.8 37.4	3.0 4.0	FL and HAZ†
Fully ferritic	Unwelded	15-20	25-35	15-25	—
	Nickel 61	19.6 19.8	27.6 27.0	11.5 11.5	Away from weld Away from weld

* Test pieces 0.564 in. dia. over 3.15 in parallel portion.
† FL—Weld fusion zone. HAZ—Heat-affected zone.

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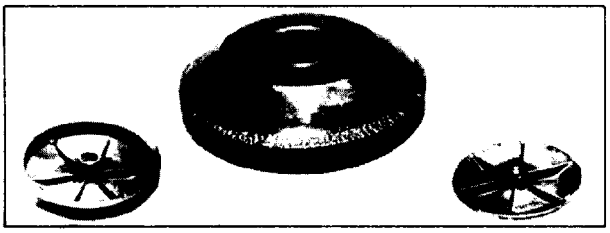


Fig. 14. *Two matched halves of a ferritic ductile iron compressor piston welded by short-arc MIG.[8] (See text for description.)*

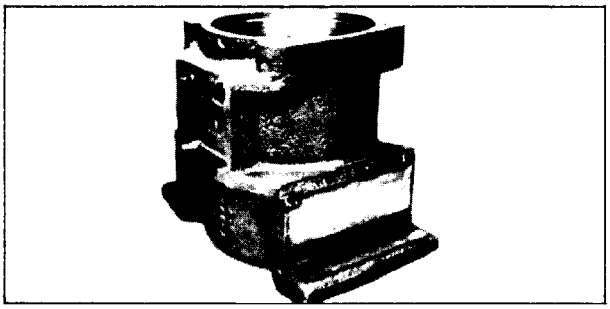


Fig. 15. *Manganese steel wear plates short-arc MIG welded onto ductile iron railway axle boxes.[8] (See text for description.)*

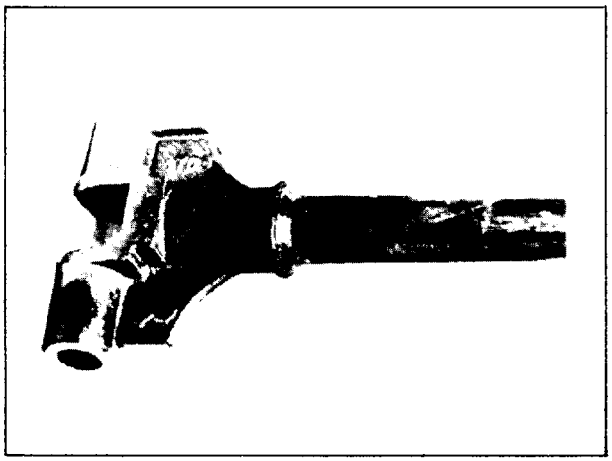


Fig. 16. *SG iron coupling welded to steel operating arm of hydraulic jack by short-arc MIG welding.[8] (See text for description.)*

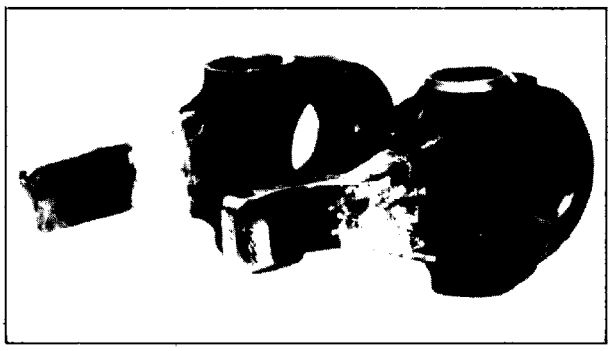


Fig. 17. *Brake housing produced by short-arc MIG welding ferritic/pearlitic ductile iron casting to steel bar.[8]*

shown in Fig. 16. In this case, short-arc welding was performed under CO_2 gas using 1 mm dia. copper-covered carbon steel filler wire. No pre- or post-welding heat treatment was performed and several thousand such hydraulic jacks have been produced successfully.

Fig. 17 shows a mild steel bar welded to a heat-treated ferritic/pearlitic ductile iron casting for the operation of a brake. The short-arc MIG process was used, using 0.8 mm dia. pure Ni wire under argon gas shield, without pre- or post-weld heating.

Another example of welding to ductile iron pipe is seen in Figs. 18 and 19. Annealed ferritic ductile iron pipes (100–1000 mm dia.) have a self-compensating device incorporating a welded bead which takes up longitudinal forces resulting from internal operating pressures up to 4200 N/m².

Referring to Fig. 19, the weld bead (2) which forms a circumferential protrusion on the spigot (3) is subjected to a small shear stress as it comes up against the split ring (4). Because of the thrust, the split (4) is pushed against the collar ring (5) and closes around the spigot (1) as the tapered outer surface of the split ring (4) is pressed against the mating surface of the collar ring (5).

Clamping bolts (6) transmit the longitudinal force to the socket (7) of the next pipe.

On-site welding is carried out with Ni/Fe coated welding rod and in-plant welding by the short-arc MIG welding using 0.8 mm dia. Ni/Fe wire. Over 30,000 m of pipe assemblies have been laid in lengths varying from 6 m to 7400 m.

Many thousands of the heavy vehicle steering box assemblies, shown being short-arc MIG welded in Fig. 20, have been produced at the rate of 300 units per week. The mild steel column guide (Fig. 21) is fillet welded to a ferritic ductile iron steering box using a fully automatic short-arc MIG system using 1.2 mm dia. pure nickel filler wire.

Recommended process techniques

Equipment required:

a) *Specialized equipment:* power source constant voltage-type MIG welding rectifier. Coupled wire feed mechanism and regulated gas shield to special welding torch.

b) *Filler:* high nickel content wire. Wire diameter: 0.80 mm. Wire type: nickel 61.

c) *Protective gas shield:* Argon

Equipment settings:

Power source settings:	100–150 A
	20–22 V
Wire feed speed:	350–450 in/min
Gas shield flow rate:	27 ft³/h

Recommended technique: Constant weave of the torch from side to side (width of about ¼ in.) throughout each pass is required. The torch should be kept moving at all times and no attempt made to build up deposit rapidly in one location. Ideally, runs should be limited to 3–8 cm at a time in staggered fashion to minimize the heat input in any one area. Interpass temperatures should be maintained at below 150°C for fully ferritic ductile iron if maximum ductility is to be achieved. For pearlitic ductile iron the workpiece temperature should be about 250–300°C.

Wherever possible, the root run should be back ground before proceeding with the filler runs so as to remove any voids or other undesirable regions in this, the most important of the deposits. While there is no flux deposit to be removed, it is advised that each run be wire brushed before depositing the next run. The number of runs required is not critical but, as a

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guide, in welding a 1 in. thick plate with a double-Vee preparation, approximately 20–25 runs will be required to complete the joint.

Gas welding

Gas welding covers (a) Gas fusion welding and (b) Powder welding.

Gas fusion welding

Gas fusion welding is a well-established welding method which has proved to be a suitable technique for welding cast iron materials including numerous successful welds of Meehanite metal. The process simply involves fusion of the base metal and filler rod by heat generated from an oxy-acetylene flame, the weld pool being constantly fluxed. Deposits of ductile iron are possible using ductile iron filler rods and fluxes of suitable composition (basically cerium or other rare earth elements required). In all gas welds, success depends upon careful and adequate preheat of the workpiece and maintenance of a controlled and well-fluxed weld pool. Using ductile iron filler rods, strength and ductility properties comparable with the base iron can be achieved.

The major drawback to gas welding is the fact that it is a time-consuming process and the amount of heat input to the casting is liable to cause serious distortion.

An example of a gas weld is shown in Fig. 22. Pearlitic ductile cast iron pump flanges 1240 mm dia. and 12 mm thick were welded in the as-cast condition using oxy-acetylene torches. Preheating to 300°C and post-weld stress relieving heat treatment were employed.

Recommended process techniques

Equipment required:

Heat source: Oxygen and acetylene.

Filler: Rods of ductile iron, nickel base alloys, etc. are commonly used.

Flux: As recommended by filler rod supplier. Generally, borax base for flake irons, rare earth/silicon base for nodular irons.

Gas torch: Standard equipment.

Equipment settings:

Oxy-acetylene: Neutral or slight excess of acetylene flame.

The size of the filler rod relates to the amount of heat required to melt it and, therefore, to the thickness of the material being welded.

Material thickness (maximum)		Rod diameter (maximum)		Acetylene consumption	
in.	mm	in.	mm	(ft^3/h)	(m^3/h)
3/8	9	1/4	6	15-30	0.4 -0.85
3/8-3/4	9-19	5/16	8	30-60	0.85-1.70
over 3/4	>19	3/8	9	60-110	1.70-3.0

Recommended technique. After preparation, the castings should be uniformly heated to 600°C if using a ductile iron rod. (If using a low melting point filler rod then preheat to 300–350°C).

Forehand welding technique, with the rod ahead of the advancing weld is advised, since this enables a better puddling action to be obtained with consequent flux scouring of inclusions and slag. Good fluid fluxing is essential in gas welding to avoid slag entrapment and maintain nodularity in ductile iron welds. The base iron joint face should be first heated with the gas torch until melting commences and then the filler rod melted by immersion in the pool, not by direct melting off in the flame. Gentle stirring of the pool is recommended as welding

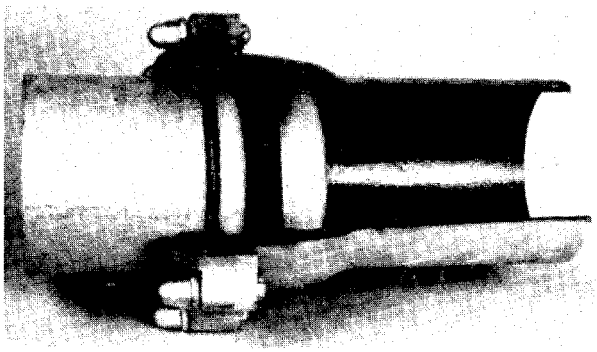


Fig. 18. *Ductile iron pipe joint with short-arc MIG welded circumferential bead.*

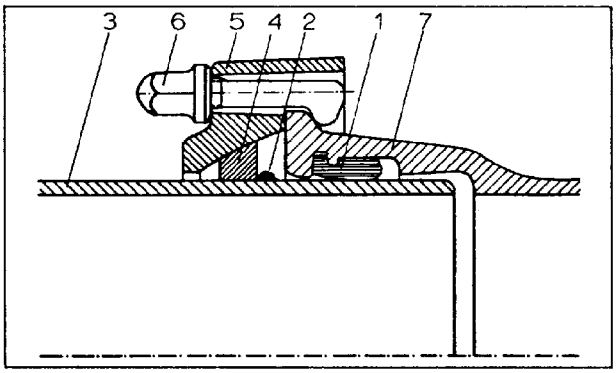


Fig. 19. *Schematic arrangement of pipe joint.[8] (See text for description.)*

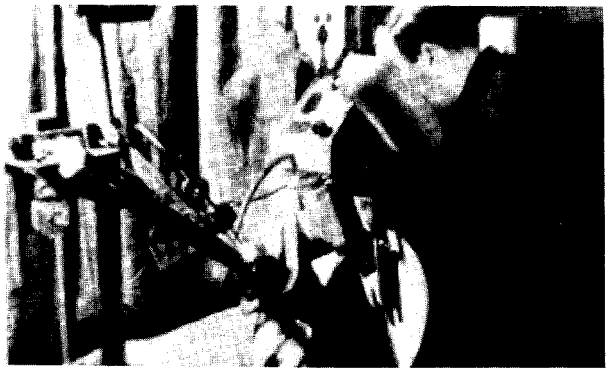


Fig. 20. *Mild steel column welded to ferritic ductile iron steering box by short-arc MIG process.[8]*

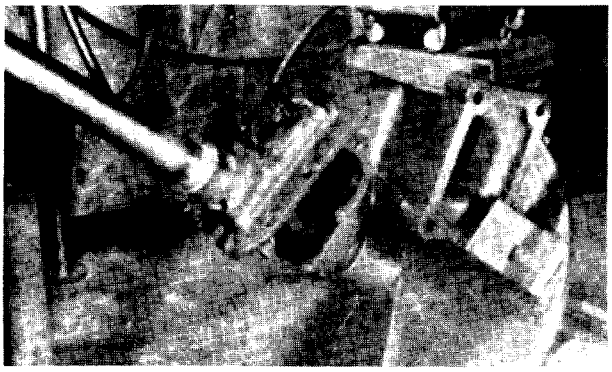


Fig. 21. *Same as Fig. 20. (See text for description.)*

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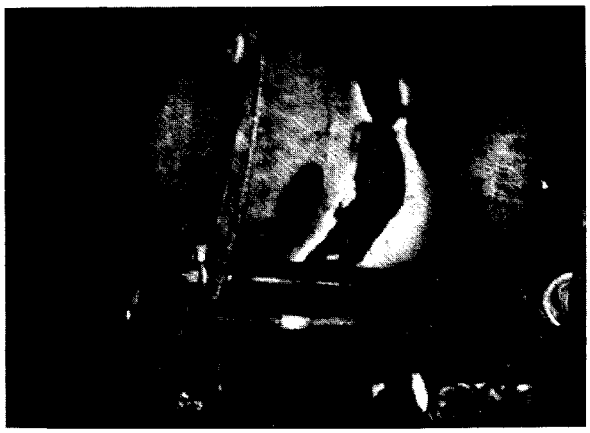


Fig. 22. *Gas fusion weld of pearlitic ductile iron pump flanges.*[5] *(See text for description.)*

proceeds with the rod being periodically dipped into the flux. After completion of the weld, slow cooling to room temperature is advised. For castings which are required to be machined, an annealing heat treatment is recommended.

Powder welding

Powder welding is a non-fusion welding technique and may simply be regarded as a low heat input gas welding process in which filler rods are replaced by powdered metal particles which are introduced into the gas stream. The process employs an oxy-acetylene torch of design (Fig. 23) which incorporates a metering valve to regulate the flow of the powdered welding alloy into the oxygen stream flowing through the torch.

As the powdered alloy particles pass through the flame, they are progressively heated prior to impinging on the base iron surface. The melting point of the deposited alloy is below that of the base iron and when the base iron surface temperature reaches a certain level, the deposited powder coating melts and

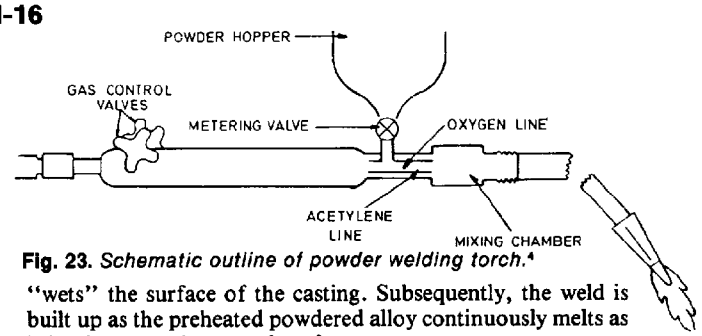


Fig. 23. *Schematic outline of powder welding torch.*[4]

"wets" the surface of the casting. Subsequently, the weld is built up as the preheated powdered alloy continuously melts as it impinges on the wetted surface.

The bond thus produced is a diffusion bond and, therefore, the success of the weld, in terms of strength, is determined by the effectiveness of the development of this diffusion bond. In this respect, a critical requirement is the necessity to achieve complete "wetting" of the base iron surface. No flux is required, but it is imperative that the surface of the base metal is completely free of all surface defects, oil, grease, etc.

From a metallurgical viewpoint, the process produces welds which are free from carbides and, while the metal in the vicinity of the weld has a heat-affected zone, this can be eliminated by suitable post-weld heat treatment.

Powder welding thus offers the following advantages to the process selector:

1. Low preheat temperatures.
2. Powder welds are relatively easy to machine and are free of carbides.
3. Capital equipment costs are low.
4. No flux is required.

The process has the disadvantage of being relatively slow compared with electric arc methods, the powder weld has virtually no ductility and this is, therefore, a limitation with ferritic base irons, if this property is required, and, finally, the cost of the powdered metal filler is relatively high. Nevertheless, there are several applications mainly for defect repair and other welds on castings where powder welding could be usefully and profitably employed.

A typical example of a powder weld repair to a small ferritic ductile iron diesel engine top plate cover is shown in Fig. 24.

Mechanical properties

In an evaluation[1] of the powder welding process with the Meehanite grades of nodular iron, base metal tensile strength in

Table 5

MECHANICAL PROPERTIES OF BUTT-WELDS MADE IN PEARLITIC AND FERRITIC S.G. IRON PLATES USING THE POWDER WELDING PROCESS[6]

Material	Tensile properties of butt-welds made with nickel-base powder alloys			Hardness, HV 30		
	Tensile strength, tons/in.²	Elongation on 2 in., %	Position of fracture	WJ	WM	BM
Pearlitic unwelded	50	5		—	—	290-304
Pearlitic as-welded	35.2 29.2	1 1	WJ/WM WJ/WM	360-410	350-380	290-304
Ferritic unwelded	27.0	25.0				148
Ferritic as-welded (Test 1)	22.7 23.3 26.9	1½ 3½ 4½	WJ/WM WJ/WM WJ/WM	235-245	210-230	148
Ferritic as-welded (Test 2)	27.6 27.6	11.0 11.5	BM BM	220-250 220-250	200-220 200-220	148 148

WM—Weld metal; BM—Base metal; WJ—Weld junction.
The nickel-base powder alloy used to weld the pearlitic nodular iron was reported to have a tensile strength of 35.0 tons/in.² and a deposited hardness of 36–42 HRC.
The nickel-base powder alloy used to weld the ferritic irons in Test 1 was reported to have a tensile strength of 33 tons/in.² and a deposited hardness of 15–20 HRC.
The powder alloy used to weld the ferritic irons in Test 2 was a copper-containing nickel alloy.

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powder welded ferritic nodular test plates was achieved but the elongation was reduced from 12% to less than 2%.

These results are in agreement with other reports[4,9] which confirm that good tensile properties are achievable with powder welding but ductility, as measured by elongation, is very poor (Tables 5 and 6).

Recommended process techniques

Equipment required:

Heat source: Oxygen and acetylene.

Filler: Powdered metal alloy—usually high in nickel content.

Torch or gun: Specially designed oxy-acetylene welding torch to allow the introduction of the powdered alloy into the flame.

Equipment settings:

Oxygen and acetylene: Employ a slight excess of acetylene in order to ensure slightly reducing flame condition.

Torch tip: If any quantity of powder welding is envisaged, a water-cooled torch tip is advised.

Recommended technique: A successful powder weld is largely dependent upon control over the preparation of the surfaces to be welded. Machined surfaces are desirable, entirely free from dirt, grease and oil.

The first layer deposited is most important. The prepared faces should be carefully heated to 200–300 °C with the welding torch and then rapidly coated with a thin film of powdered alloy. This initial layer acts as a protection against oxidation during the main build-up of the deposit. The subsequent build-up can then proceed continuously. Only if there is a likelihood of overheating occurring with a heavy deposit should multi-pass welding be conducted.

The weldment should be allowed to cool slowly to room temperature preferably in an oven or in an insulating material.

Conclusions

1. The welding of ductile iron castings, although presenting greater difficulties than mild steel or certain malleable irons, is quite possible.
2. An understanding of the metallurgical factors involved, particularly those associated with the high carbon content of ductile irons, assists in developing and applying the necessary process controls to avoid or minimize the dangers of carbides, cracking, hard spots, etc. during welding.

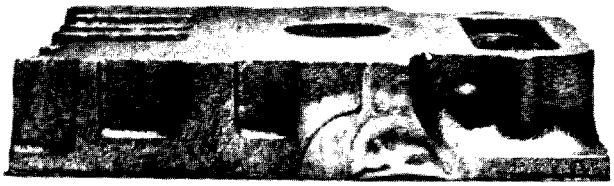


Fig. 24. *Powder weld repair[1] of ferritic ductile iron cylinder head.*

3. With correct selection of welding process and filler rods/wires, all grades of ductile iron—ferritic, pearlitic, austenitic—are weldable.
4. Of the processes discussed, short-arc MIG welding provides the best possibilities of achieving satisfactory welds with good mechanical properties. Nevertheless there are many examples where functionally satisfactory welds are regularly achieved with metal arc, gas fusion and powder welding processes.
5. In all cases, successful welds require careful attention to preparation including cleaning, grinding, joint design, etc.
6. No weld can match the base metal in all respects. Some compromise is always involved whether it be in mechanical properties, physical properties, color match, etc. The object is to produce a joint or a repair capable of enabling the component or assembly to fulfill its function.
7. It is hoped that this paper will serve to encourage foundrymen, designers and users of ductile iron castings to consider welding as a means of extending further the usefulness and application of nodular iron. □

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Acknowledgements

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Welding of Malleable Cast Iron

CIATF Commission 7.2
Malleable Cast Iron
Chairman: Dipl.-Ing. H. G. Trapp
Schaffhausen, Switzerland
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Dusseldorf, Federal Republic of Germany

This is the Final Report of the CIATF Malleable Cast Iron Commission 7.2. The *ICMJ* is pleased to publish it in an unabridged English translation.

INTRODUCTION

The broadsheet N 70 of the Verein Deutscher Giessereifachleute was taken as a basis for the first part of the report. The examples of applications were made available by the following companies: Bergische Stahlindustrie, Remscheid, and Georg Fischer AG, Schaffhausen, as well as by the Centre Technique des Industrie de la Fonderie, Paris. They were arranged and described by Ing. (grad.) P. Tolke, Schaffhausen.

DEFINITION

Welding Varieties

The following weldings of castings are distinguished:[1]

1) production welding, which is the welding undertaken by the producer of castings in the course of manufacturing with the aim in view of the guaranteed properties and to secure the requisite quality of castings;
2) construction welding for joining castings mutually or with other structural members; and,
3) maintenance welding (repair welding) for the repair of castings, which are damaged during their applications. In this case the castings are restored to suitable properties and usability by removing defects.

Weld Quality Grades

Two quality grades are distinguished:[2]

1) Quality grade A: The weld has the same properties as the unwelded material; and,
2) Quality grade B: The weld has different properties as the unwelded material, but the weld quality suffices the requirements for a special purpose (conditional quality).

MALLEABLE CAST IRON

Classification

According to ISO DIS 5922, two kinds of malleable cast iron can be distinguished.

1) blackheart (and pearlitic) malleable cast iron of the grades B 30-06 to P 80-01; and,
2) whiteheart malleable cast iron of the grades W 35-04 to W 45-07.

Among them W 38-12 has a special place for construction welding requiring a higher strength in the welding joint.

Microstructure and Carbon Content

Blackheart malleable cast iron posesses a microstructure which is in practice not influenced by the section thickness. For the grades B 30-06 to B 35-10, the microstructure consists of ferrite and temper carbon; for the grades P 45-06 to P 80-01, it consists of a matrix with an increasing content in combined carbon and temper carbon. The total carbon content lies over 2% for this group of materials.

Decarburized (whiteheart) malleable cast iron is characterized by the decreasing carbon content from the core to the surface of the casting, which results in different microstructures in various zones. The core zone consists of pearlite and temper carbon with a small proportion of ferrite, the intermediate zone of pearlite and ferrite with a residue of temper carbon, and the surface zone of ferrite. The total carbon content in the core zone lies over 1.5% C, and in the surface area below 0.1% C.

Weldable malleable cast iron is intensively decarburized. Carbon content is one of the main factors which affects the

Source: AFS International Cast Metals Journal, September 1982 (American Foundrymen's Society)

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suitability for welding. The carbon content is restricted to ≤ 0.3% for section thicknesses ≤ 8 mm (see paragraph below).

SUITABILITY FOR WELDING

With regard to the suitability for welding, two groups of malleable cast iron can be distinguished.

Weldable Malleable Cast Iron

Without additional expense before, during or after welding with common unalloyed filler metals, quality grade A can be obtained.

The pre-condition for this is that the carbon content in the weld area does not exceed the approximate value of 0.3%. This condition can be reached by an intensive decarburizing annealing up to a section thickness of max. 8 mm.

The chemical composition of malleable cast irons is not standardized. Weldable malleable cast iron is produced with variable chemical composition:

W 38-12 with a composition adjusted for weldability (Si≤ 0.55%, S ≤ 0.15%, Mn approx. 0.8%).

This grade of material is suitable for all welding processes without restriction and can be used for welding joints which must meet an acceptance test with an internal pressure load.

Materials which were cast with a common chemical composition of whiteheart or blackheart malleable cast iron have been proved weldable by mechanized or automatic welding processes such as resistance flash welding, friction welding, and gas metal arc welding, as well as manual arc welding with basic electrodes.

Quality grade A is also obtainable with normal carburized whiteheart cast iron of grades W 35-04, W 40-05 and W 45-07, if the section thickness does not exceed 4 mm and the material control in respect of weldability is ensured. The weldability of fittings is not guaranteed due to the decarburization degree adapted for an optimal machinability.

Whiteheart and Blackheart Malleable Cast Irons Which Are Conditionally Weldable at Enhanced Expense

The microstructure of high carbon malleable cast iron contains, besides varying proportions of combined carbon, varying proportions of free carbon (temper carbon). Without additional expense, welding with dissimilar filler metal produces only quality grade B. If quality grade A is demanded, the welding of whiteheart malleable cast iron with a carbon content of over 0.3% (in the weld) must be preceeded by a preheating of up to at least 250 to 400C. If the carbon content is so high that, in spite of preheating, ledeburite is formed, then a one-stage or two-stage annealing is required after welding. Blackheart malleable cast iron demands basically a suitable postweld heat treatment. According to experiences, the graphitizing products of ledeburite produced by welding lead to properties which do not correspond to that of the unwelded material. The Ni-martensite zone in the weld of malleable cast iron/filler wire, being unavoidable in welding with a high nickel filler metal, cannot be eliminated even with a high temperature heat treatment.

WELDING TECHNOLOGY

The material and quality grade determine the necessary welding technology.

Edge Preparation

As far as possible, only blank metallic surfaces should be welded. In construction welding, the skins of castings should be thoroughly removed so that the impairment of the welding process and/or weld metal can be eliminated.

In production welding, the joint area must be finished down to the sound base material.

The surface zone of whiteheart and weldable malleable cast iron grades consists of an oxide layer, the so-called temper skin. The oxides can react with carbon in the base metal and/or filler metal to form carbon monoxide and produce pores in the weld. This holds true for all fusion welding processes, because the temper oxides easily will come into contact with the weld pool. The CO-pores are readily avoided by removing a maximum of 0.3 mm temper skin in the welding area. If this is not possible, i. e. the temper skin must be welded, the defect formation can be extensively reduced by choosing suitable filler metals (lime-covered electrodes, sufficient deoxidation effect). On a limited scale, defect formation can be reduced by choosing welding processes (pressure welding with reinforcement) as well as by adjusting the torch.

Edge preparation can be chosen corresponding to the requirements of the welding process, as in the case of welding steel. In resistance pressure welding and pressure welding, it must be taken into consideration that the different melting points in the welds of malleable cast iron/steel results in more melting loss and thicker protrusions.

Heat Introduction and Heat Treatment

In cold welding without postweld heat treatment, the weldable malleable cast iron within the composition range described above produces quality grade A. For other malleable cast iron grades, the quality grade of the welds is fixed by the manner of heat introduction before, during and after welding as well as the type of postweld heat treatment. Cold welding, i. e. welding without or only with a slight preheating, leads to quality grade B with the normal malleable cast iron grades. The properties can be improved by one or more stages of heat treatment after welding, for which a tempering of martensite is usually sufficient. A prior graphitizing annealing normally destroys the ledeburite, and thus improves the machinability, but results in lamellar graphite products in the area of ledeburite isles formed during welding, which can be associated with a decrease of elongation and fatigue strength.

If quality grade A is demanded on malleable cast iron grades with higher carbon contents, hot welding with a pre-heating to temperatures in the range of 250 to 400C (according to the complexity of the casting) must be applied. Parts of whiteheart malleable cast iron which have been decarburized to ≤ 2% C, i. e. under the limit for ledeburite formation, require no additional graphitizing heat treatment. This graphitizing heat treating must be carried out, however, on parts of blackheart malleable cast iron. Semi-hot welding with a localized preheating up

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to 250 to 400C can reach quality grade A for parts of whiteheart malleable cast iron which have no potential for ledeburite formation (C ≤ 2%).

Welding Processes

Welding Processes for Weldable Malleable Cast Iron — The chemical composition of W 38-12, being adapted for the best weldability and combined with an intensive decarburization, enables production, construction and repair welding of these materials with all known welding processes and filler metals not requiring a postweld heat treatment, if the section thickness does not exceed approximately 8 mm. Recent developments[3,4] have shown that malleable cast iron grades with a composition deviating from grade W 38-12 (for example, whiteheart or blackheart malleable cast iron) are also suitable for construction welding if the carbon content can be reduced to ≤ 0.3%, by an intensive decarburization. In the field of quantity production, these materials are preferably welded by mechanized and automatic welding processes such as gas metal-arc welding with carbon dioxide, argon and gas mixture, flash welding, friction welding, and for small lots metal-arc welding with covered electrodes. Diffusion welding, plasma-arc welding and seam welding have been also tested with success.

Welding Processes for Whiteheart Malleable Cast Iron — Practically all welding processes are possible. The requisite expense for obtaining a sufficient weld grade is confined to the application of manual metal-arc welding in small lot production. Free temper carbon can result in CO-gas pores even in metal-arc welding conditions. Therefore, gas welding should only be applied with a reducing flame.

Welding Processes for Blackheart Malleable Cast Iron — If production welds are possible, the metal-arc welding process with covered electrodes is preferable. The welds should have a heat input as low as possible, that is to say, electrodes with a small diameter, with a welding current as low as possible, in order to reduce the hardening effect (martensite and ledeburite). Additionally, heat input can be reduced by making thin, short weld beads. The length of the weld bead should not exceed 10 times the electrode diameter.

Under defined strain conditions, welds with the MAG welding process are applied to P 80-01 grade malleable cast iron on the production line.

Welding Filler Metals

Welding Filler Metals for Weldable Malleable Cast Iron — For weldable malleable cast iron, unalloyed weld filler metals can usually be applied. Materials having a composition within the range of W 38-12 chemical composition can be welded either with iron ore and rutile or lime-covered electrodes.

For gas welding of W 38-12, gas welding rods or wires can be applied.

Gas metal-arc welding of all grades of weldable malleable cast irons, as offered on the market, can be done with unalloyed or flux cored wire. For weldable materials on a base of whiteheart or blackheart malleable cast irons (see above), basic covered electrodes should be applied in any case. In all cases, quality grade A will be obtained.

Weld Filler Metals for Whiteheart and Blackheart Malleable Cast Iron — For whiteheart malleable cast iron grades with high carbon content and for blackheart grades, both similar and dissimilar weld filler metals are applied. There are covered or bare electrodes as well as welding rods on the market.[5] The classification of covered electrodes is based on the chemical composition of the stick. The weld properties depend also on the stick composition, mainly on the coating, on the alloying from the coating and base metal, on the welding process, the welding parameters, and on the cooling rate. Uncovered electrodes and welding rods are also classified according to their composition. Their application is only recommended for hot and semi-hot welding. Weld filler metals which produce a weld with lamellar graphite are not suitable for the welding of malleable cast iron.

QUALITY ASSURANCE

Construction Welding

For construction welding, weldable malleable cast iron shall have priority. The required actions for quality assurance must be coordinated with safety requirements and the use of the cast construction.

The implementation of the following quality assurance actions must be considered accordingly:

1) required proof of the qualifications of the factories;
2) examination of integrity and preparation of the following working papers;
   a) calculation of strength
   b) operational drawings
   c) guide on the preweld or postweld treatment and examination
   d) welding instruction, welding scheme
   e) weld testing scheme;
3) selection of suitable base metals (material permission);
4) selection of suitable filler metals and auxiliary materials (permission);
5) correct weld construction with regard to load conditions, material behavior, suitability for welding, finishing conditions, weld safety;
6) special actions for welding production:
   a) selection of welding processes (accepted for the application purpose)
   b) weld edge preparation
   c) welding jobs performed by certified welders under supervised conditions (welding inspection)
   d) right selection of operational data, diameter of the filler metal, polarity, welding current, weld buildup, welding position, compliance with the welding plan, preheating, etc.
   e) appropriate refinishing
7) Testing procedures in compliance with the requirements for the test categories (destructive and nondestructive), including documentation; and,
8) acceptance test procedure.

Production and Repair Welding

Because of the associated metallurgical factors, namely high carbon content in the matrix and the presence of free graphite under certain circumstances (except for thin sections of

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whiteheart and weldable cast iron), suitable measures for safeguarding weld quality are demanded.

Production welding for castings requiring a certificate need the agreement of both client and producer.

Technical Hints

Welding Filler Metal — The welding filler metal must match with the base metal.

Welding Process and Welding Technology — Welding process and welding technology must be coordinated with the material microstructure in the joint.

Welders — The welders must receive a sufficient training in the welding of corresponding materials.

Welding Conditions — For perfect welding performance, the following factors are necessary: selection of working conditions, size of welding filler metals, polarity, welding current, weld seam buildup, welding position, heat introduction and the application of suitable working materials, fixtures and tools. Fixing these data in a working guide is recommended.

Heat Treatment after Welding — A suitable postweld heat treatment is required for the welding of castings whose carbon content exceeds 0.3%.

Testing Methods — For proof of their soundness, welds can be tested by a suitable testing method, for example, hardness test, magnetic particle test, selective metallographic test. Remember that for the magnetic particle test a false indication can occur due to hardening in the heat-affected zone or the application of Ni-base welding filler metal.

EXAMPLES OF APPLICATIONS

There are many examples of weldments in which one or more of the parts are malleable iron castings. Because malleable cast iron is a typical mass production material, these examples often concern mass produced welded joints made by mechanized or automated welding equipment. Incorporating the welding process into a machining center does not permit any pre- or post-heat treatment of the weld region. Due to this, together with the high strength and safety requirements, the use of weldable whiteheart malleable cast iron is preferred in the automotive industry. This is a material which does not require any additional treatment — neither before, nor during, nor after welding — and gives class A welds using the normal unalloyed welding rod.

The pictorial examples are shown in the following order: Automobile production/automotive industry drive system, steering, axle parts, chassis, exhaust system; plumbing and heating industry; material transportation industry; agricultural machine industry; general machinery industry.

Figure 1 shows various designs of drive shaft flanges. As shown in Fig. 2, they are welded to the tubular steel part of the drive shaft by friction welding or by the active-gas metal-arc welding method (MAGC) under a carbon dioxide atmosphere. The condition of the drive shaft is very important for a well-functioning vehicle. Accordingly, the weld quality and weld strength properties must conform to high standards to guarantee a highly reliable drive system.



Fig. 1. Drive shaft flanges (material: blackheart malleable cast iron) which are welded to the tubular steel shaft part of the drive shaft. Weight: 0.8 to 1.2 kg.

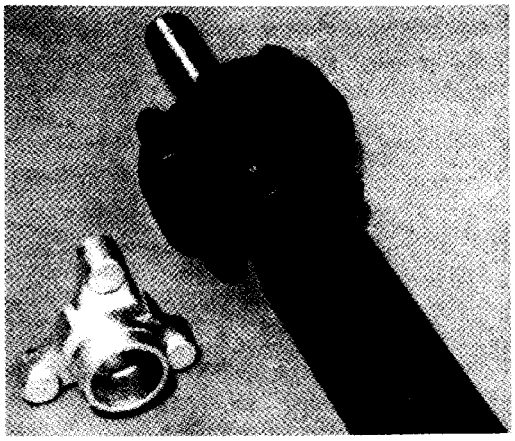


Fig. 2. Drive shaft flange of blackheart malleable cast iron welded to tubular steel drive shaft.

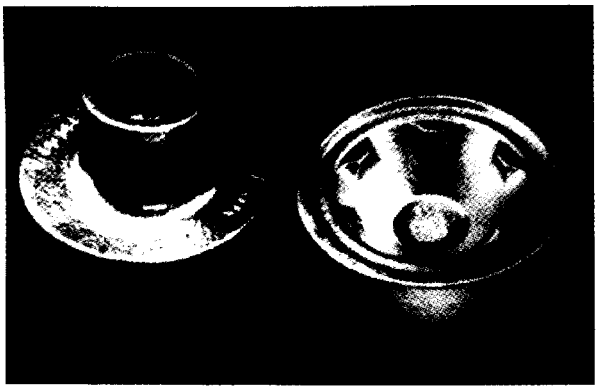


Fig. 3. Flange center bearing for the drive shaft of a higher-powered automobile. Material: blackheart malleable cast iron. Weight: 1.24 kg.

This same reliability is demanded of the weld between the flange center bearing and the tubular steel shaft, as in Fig. 3. The flange center bearing belongs to the drive shaft of a higher-powered 3-liter medium price automobile.

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From a technical standpoint, the drive shaft of the lower-price German automobile in Fig. 4 is noteworthy. The fork and fork flange are made of P 70-02 malleable cast iron. Integrated into the normal production process, the material and welding method (MAGC, carbon dioxide protective gas) give a class B weld. This means that because the weld is made without preheating, hardened structures (martensite, ledeburite) can occur in the microstructure. However, this design has been used in several million automobiles without any known service failures.

The steering system also has strict reliability requirements. The steering gear housing of Fig. 5 is welded to the tubular steel steering column. Naturally, the steering gear housing and weld must provide the same reliability (material properties, weldability) required of the safety part.

Axle parts are subjected to extreme loading. Thus, the special safety aspects of castings welded to each other to steel must be considered.

The welded joint of the W 38-12 weldable malleable cast iron wheel carrier has proven itself millions of times over. This casting is welded to a plate steel assembly and the entire unit comprises a semi-trailing arm rear axle for high-priced automobiles. Figure 6 shows the design for a mass-produced sedan and Fig. 7 shows the design for a fast sport car made in small series production. Fracture of the weld would have catastrophic consequences. Accordingly, the semi-trailing arm rear axle is classified as a safety part.

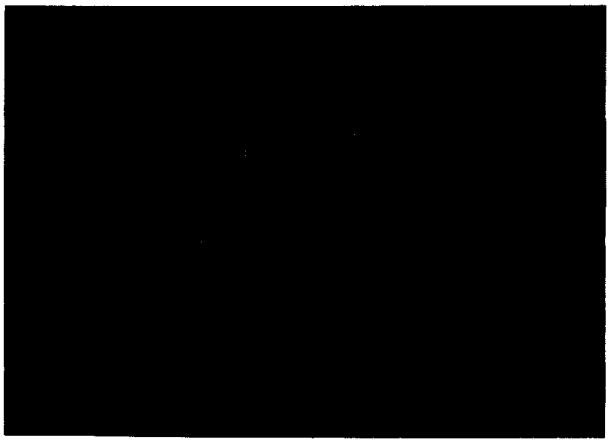


Fig. 4. Automobile drive shaft welded to universal joint.

The rear axle control arm, Figs. 8 and 9, of the legendary VW "Beetle" illustrates a special feature of welding technology. In the region of the welded joint, the W 38-12 weldable malleable cast iron wheel carrier has a wall thickness of 16 mm. This is too thick to guarantee a through-decarburization to 0.3% maximum retained carbon without uneconomical malleabilizing times. Thus, a minimum depth of decarburization, corresponding to MIG weld penetration depth, is specified for the wheel carrier to prevent hardening. In the foundry, the casting decarburization must be precisely controlled.

The rear axle segment in Fig. 10 is also from the VW "Beetle" and shows the original design. The welded assembly consists of 3 castings and 2 tubular steel segments. All five parts were welded together by means of butt welding. With further technical development, the castings of this assembly were changed so that their present-day shape is different than shown in Fig. 10. However, the welded assembly technique has remained the same.

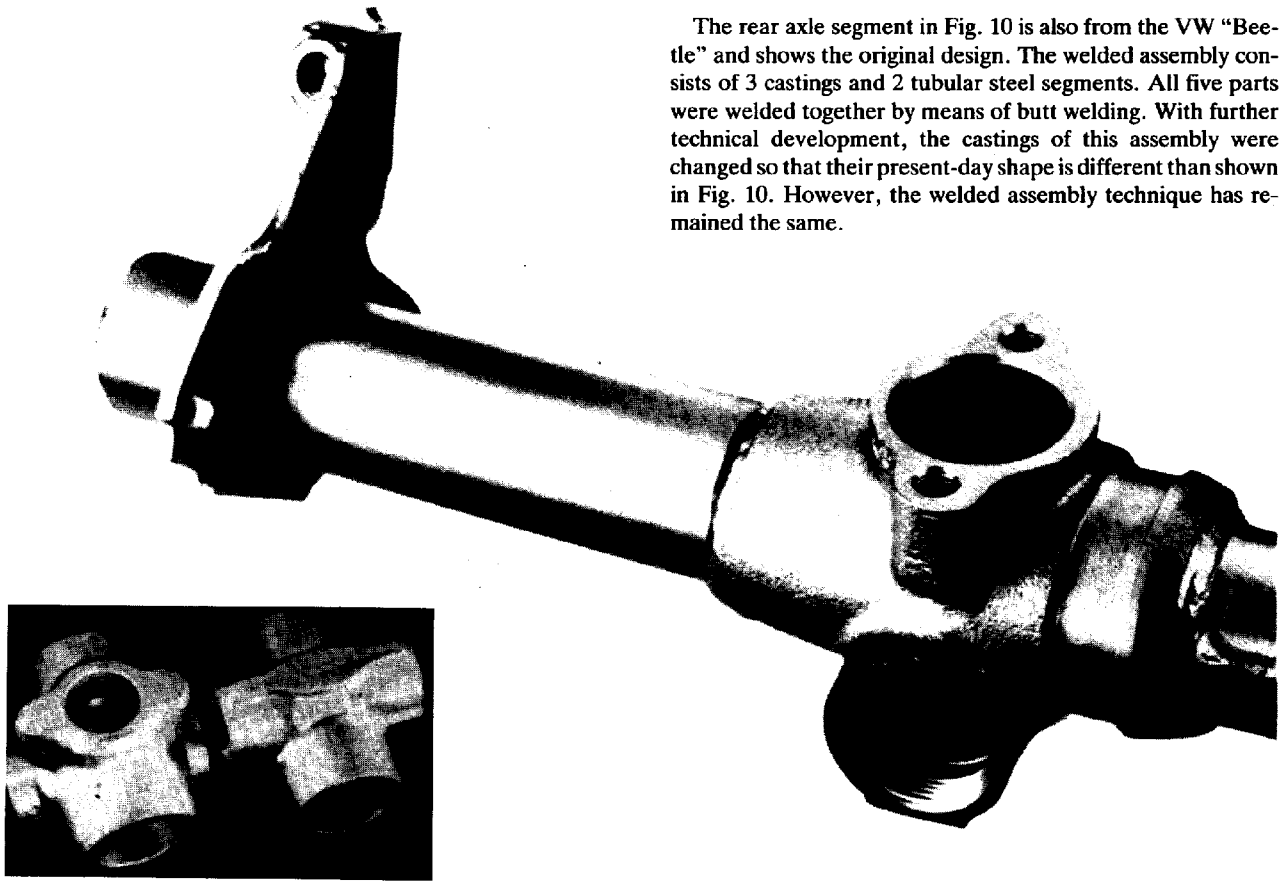


Fig. 5. Steel steering column tube with steering gear housing of whiteheart malleable cast iron. (Pref. W 38-12). Weight: 1.15 kg.

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Fig. 6. Semi-trailing arm rear axle of W 38-12 weldable malleable cast iron welded to plate steel. Weight: 3.73 kg.

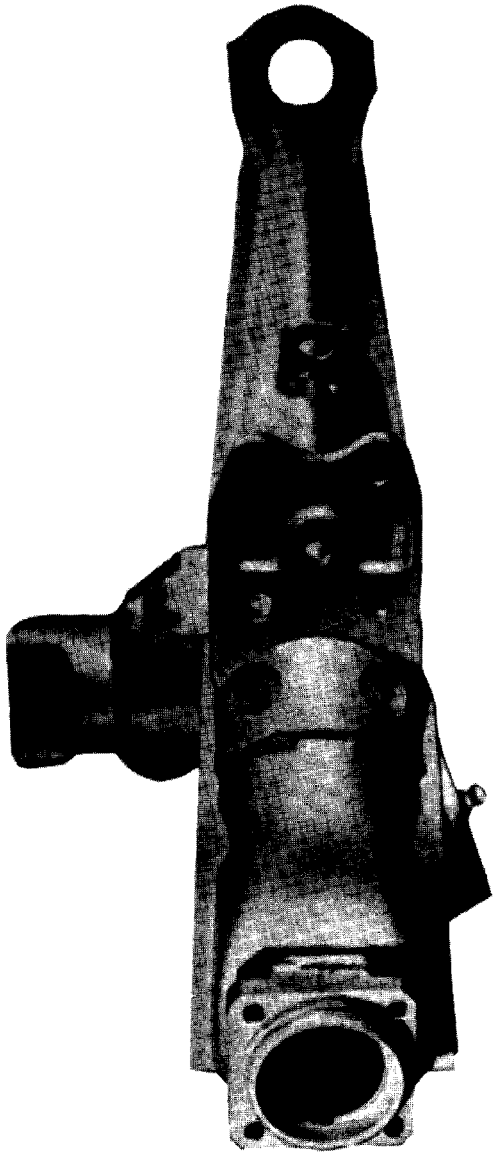


Fig. 8. Rear axle control arm.

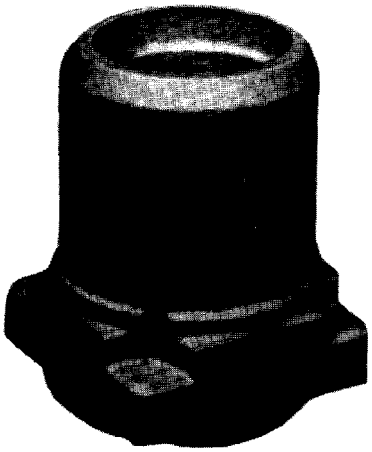


Fig. 9. Rear axle wheel carrier that is welded to control arm shown in Fig. 8.

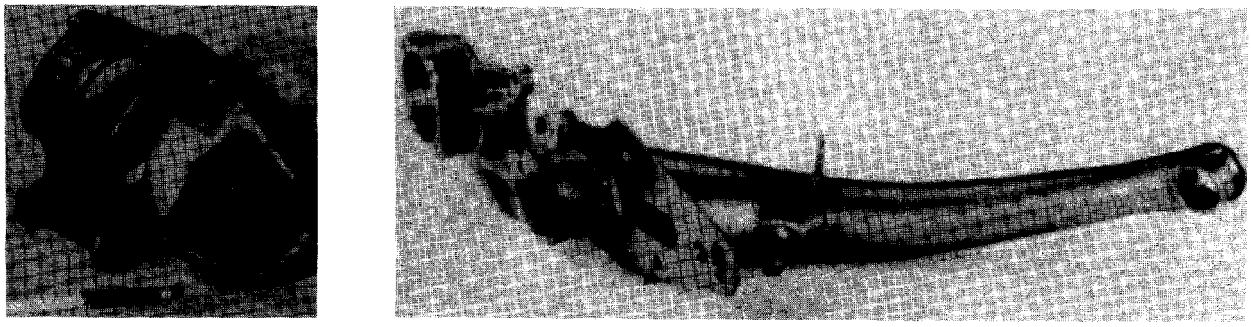


Fig. 7. Semi-trailing arm for a sport car rear axle welded to a bearing body of W 38-12 with a weight of 2.9 kg.

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Fig. 14. Shock absorber bracket, 1.97 kg. Material: Weldable blackheart malleable iron.

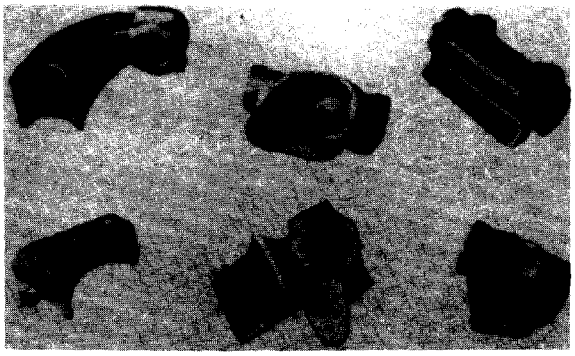


Fig. 16. Motorcycle frame spring bracket, various versions. Material: W 38-12.

By changing the pattern parting line location, more precision at the weld region was achieved. After redesigning, the part was 20% lighter For the changeover from cast steel to whiteheart weldable malleable cast iron, neither the welding method nor welding rod analysis had to be changed. Figures 15b and 15c show how the bearing functions.

The motorcycle frame spring bracket in Fig. 16 is also an interesting example of a weldable malleable cast iron application. This part has many design shapes and is welded by the carbon dioxide — MIG method or hand arc welded using coated consumable electrodes.

In Fig. 17 the cast hinge bearing (which also serves to fasten the truck bed cover) of a platform truck tailgate hinge has replaced a steel part which required complicated production op-

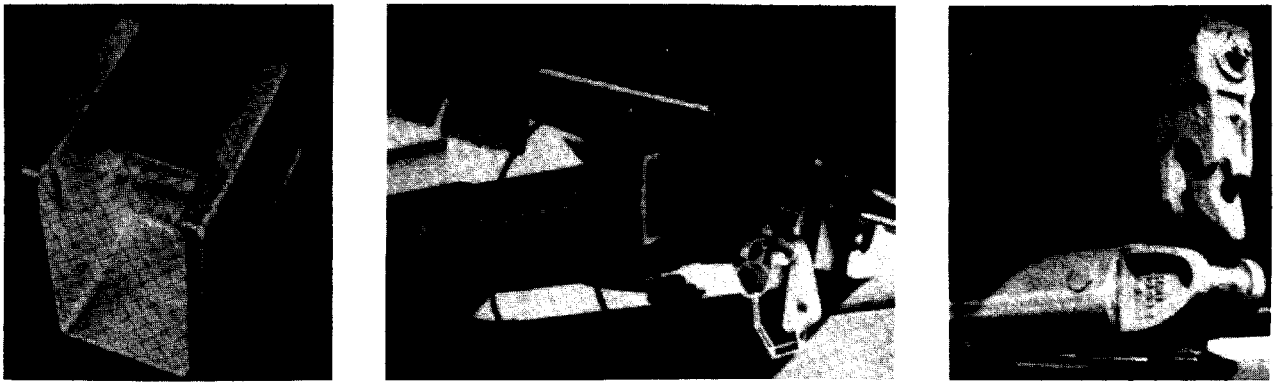


Fig. 15. Pivot bearing of weldable whiteheart malleable cast iron (1.16 kg) for a dump truck which tips to the back and to the sides; in the middle and on the right the bearing is under load in a tipped position.

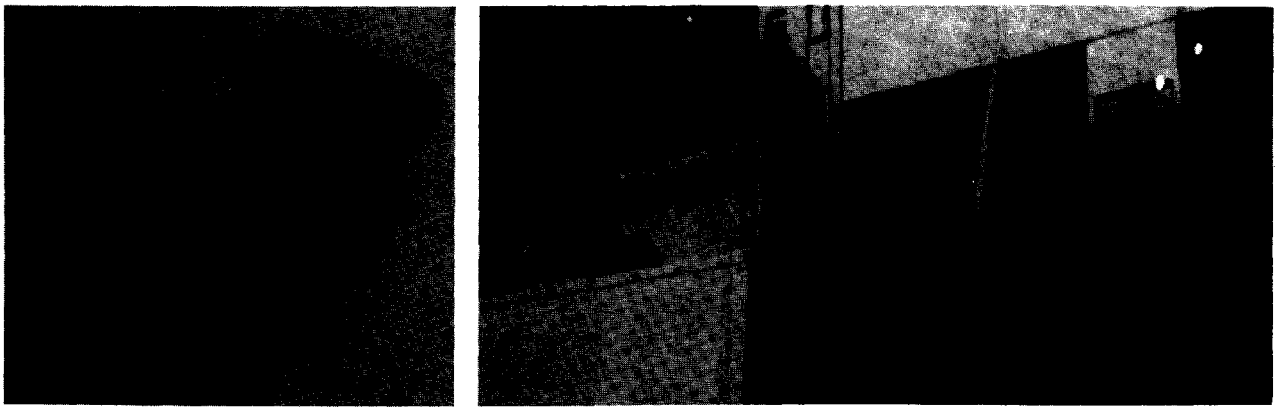


Fig. 17. Three-part tailgate hinge bearing with hoop for fastening down the truck bed cover. Material: W 38-12.

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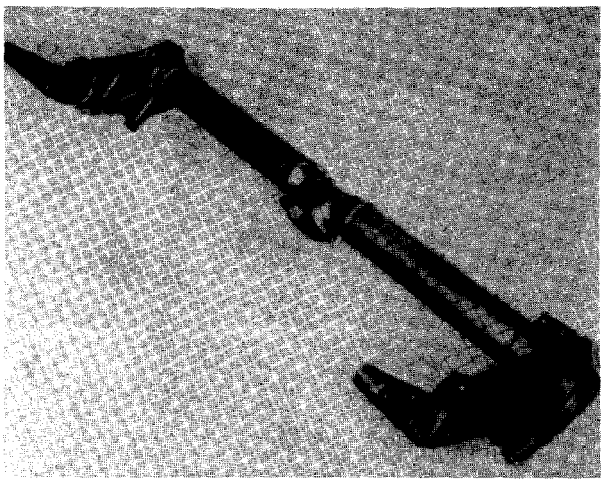


Fig. 10. Rear axle of the VW "Beetle" (in the late fifties) including two bearing bodies (3.9 kg each) and a torsion spring bearing (2.6 kg). Material: W 38-12.

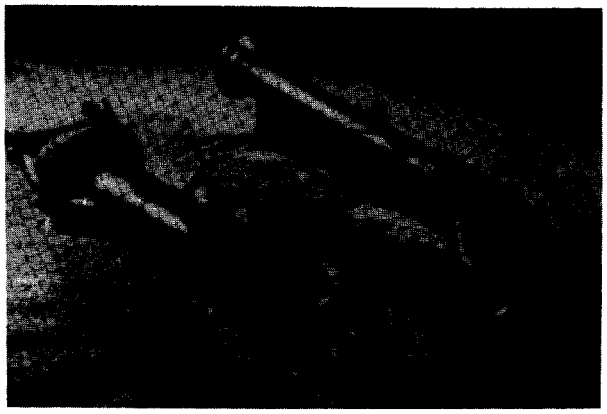


Fig. 11. Half-axle of an automobile swing axle, each including two castings of W 38-12. Weight of the forked piece: 5.5 kg.

Almost as classic and, at the time, as revolutionary an example, was the swing axle installed in a mass-produced sedan. This consisted of 4 castings butt welded to tubular steel. Figure 11 shows the half axles.

The truck planetary gear housing of Fig. 12 must be oil-tight. Thus, the MAGC weld must be free of porosity or cracks. The planetary carrier is made of weldable malleable cast iron.

From the chassis region is a bearing support shown in Fig. 13, and a shock absorber bracket in Fig. 14.

The pivot bearings of a dump truck dumper body, shown in Fig. 15, are highly-stressed safety parts. These bearings are welded to the four outriggers of the dumper body frame and with side dumping (this dump truck can dump to the back or to the left or right sides) they take up the supporting forces for the dumper body. These parts, previously made of cast steel, were redesigned to the requirements of the decarburization process.



Fig. 12. Oil-tight planetary gear housing for a truck with malleable iron carrier of W 38-12. Weight: 18 kg.

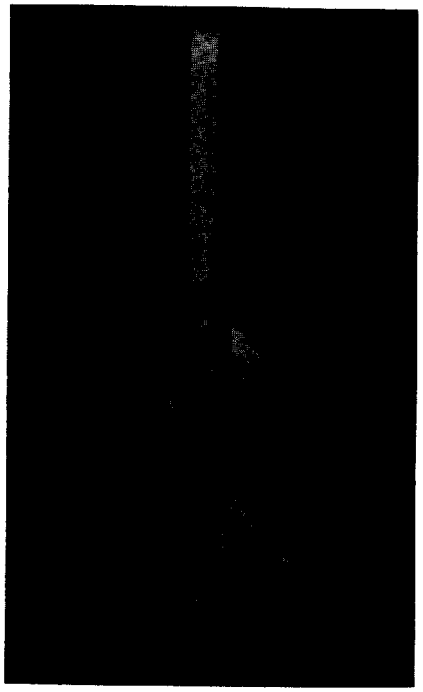


Fig. 13. Bearing support of weldable whiteheart malleable cast iron for the chassis of a passenger car. Weight: 0.46 kg.

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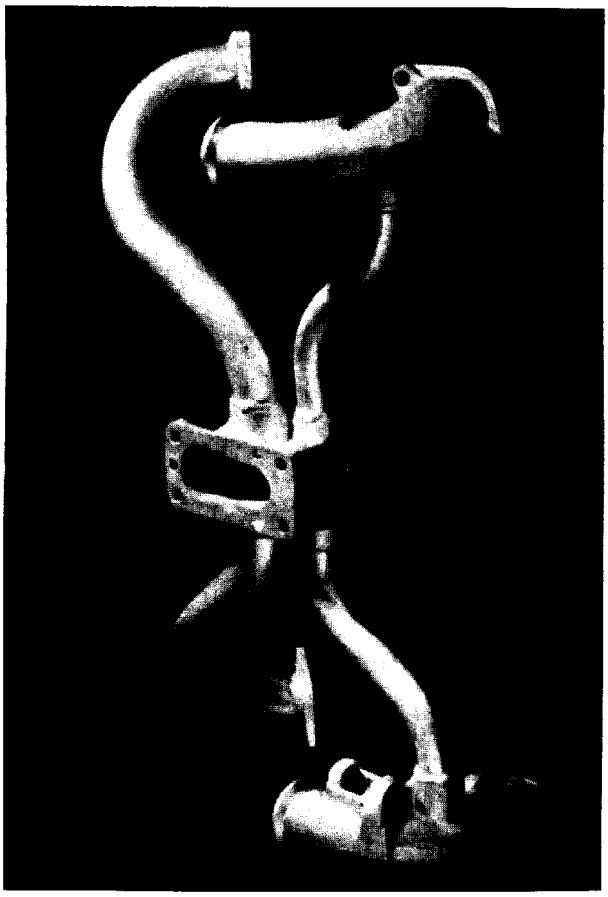


Fig. 18. Exhaust system of malleable iron castings welded to tubular steel connecting pieces. Weight: 2.12 kg.

erations including stamping, shearing, cold forming and welding. The hoop is now also cast. The spring steel clamping sleeve, which carries the hoop, is welded onto the hinge bearing.

The next examples concern exhaust systems. Figure 18 shows the complete exhaust system of a small French automobile. This consists of three malleable iron castings and four tubular steel connection pieces. The housings are subjected to high temperature stresses, as are the elbows in Fig. 19. These elbows are cast in various shapes and, by combining castings with tubular steel segments, exhaust collection systems of any desired size (such as for a 4-cylinder or 6-cylinder engine) can be constructed. Figures 20 and 21 show examples.

Another interesting example is the forked support in Fig. 22. This part is cast with a blackheart malleable cast iron analysis and then decarburized for weldability. During assembly it is held to the mating structure by three spot welds and then the main joint is made by brazing. The strong decarburization, however, guarantees that for brazing above the austenite transformation temperature (Ac_1) no sorbitic hardening occurs.

The next examples are from the plumbing and heating industry.

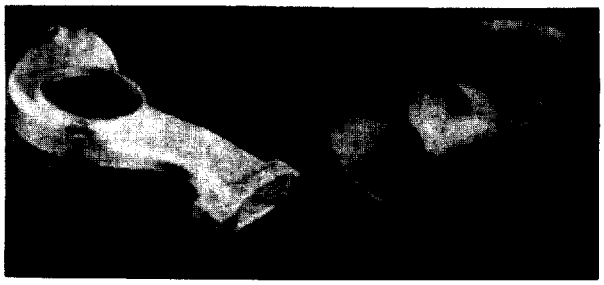


Fig. 19. Exhaust manifolds (1.6 and 2.4 kg), welded with tubular steel to an exhaust collection system. Material: whiteheart malleable cast iron.

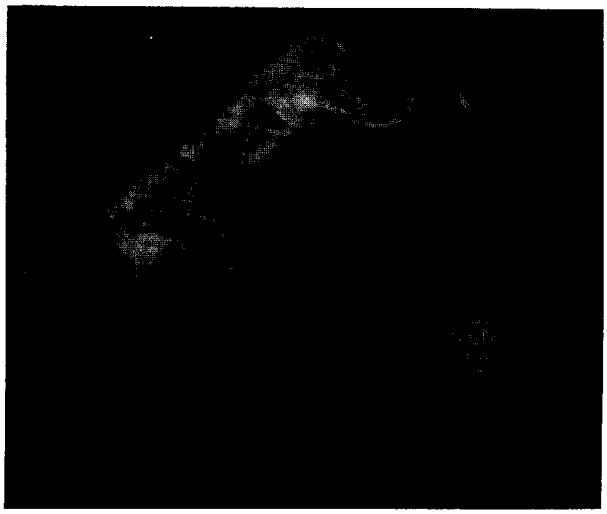


Fig. 20. Exhaust collection consisting of sheet steel and two pieces of malleable cast iron. Material: whiteheart malleable cast iron welded together.

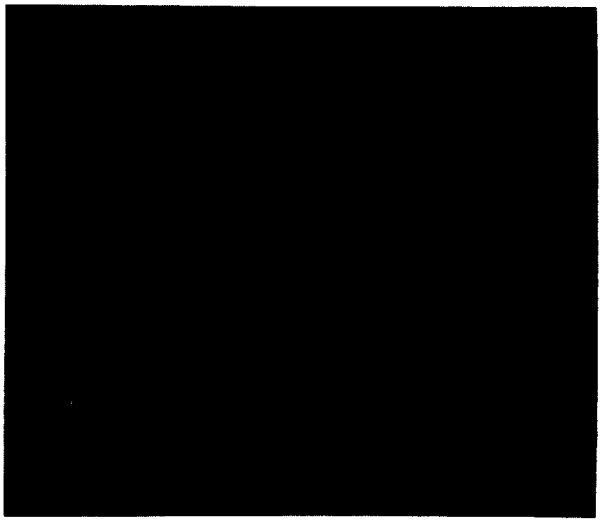


Fig. 21. Exhaust system for a six-cylinder engine constructed of the same materials as in Figs. 18, 19 and 20.

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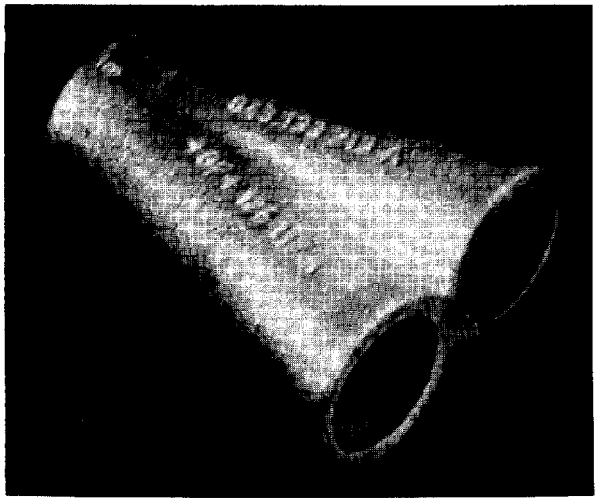


Fig. 22. Forked support for an exhaust system. Weight: 0.46 kg. Material: blackheart malleable cast iron.

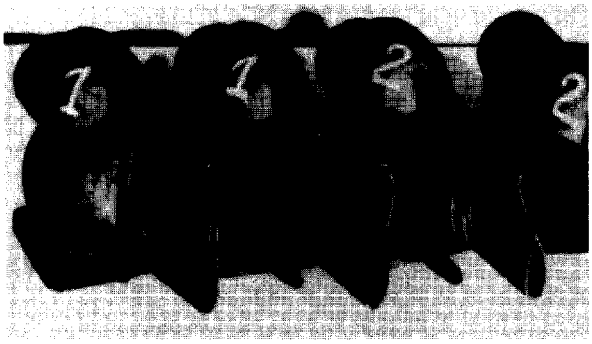


Fig. 24. Examination of the junction of weld-on valve housings shown in Fig. 23. Weld-on flange, weld and pipe endure this large deformation without any crack or fracture occuring.



Fig. 25. Radiator connection pieces of W 38-12 are produced in many different shapes. Weight of 0.2 to 0.9 kg.



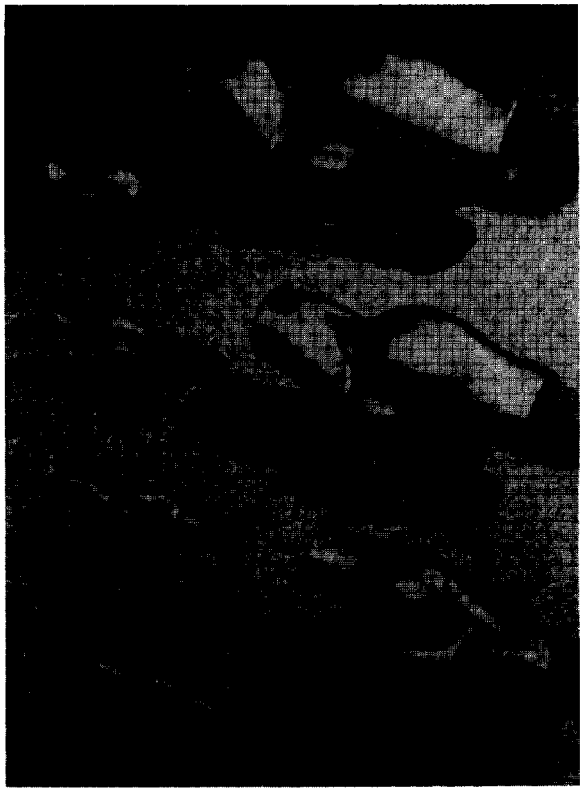


Fig. 23. Weld-on valve housings up to 200-mm nominal diameter are constructed of W 38-12. They are licensed in accordance with VdTUV (technical supervision associations) at temperatures up to 300C (see Fig. 24).

Weld-on valves are made in various sizes, Fig. 23. The high degree of deformability of the weld, pipe, and weld-on flange is seen in Fig. 24.

The radiator connection pieces in Fig. 25 are also typical applications of weldable malleable cast iron.

The casting shown in Fig. 26 is specified for a support for a hot air generator with a 40,000 kcal/h (167,500 kj/h) capacity. This heater is used to heat coaches and diesel locomotives of

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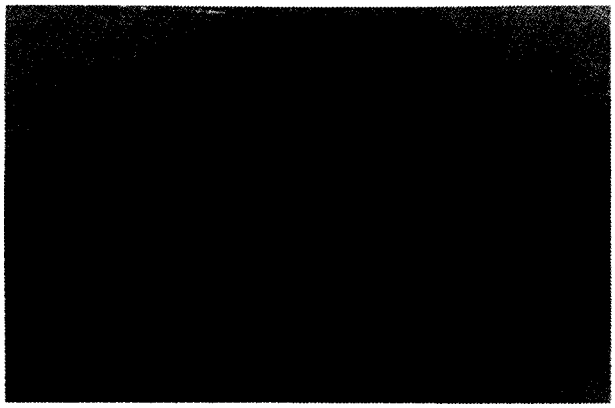




Fig. 26. Housing for a hot air generator. Material: P 45-06 welded to a steel tube.

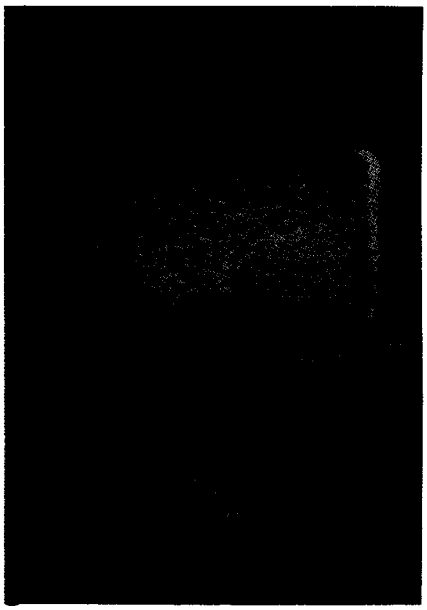


Fig. 27. Base for a sheet frame container. Weight: 0.53 kg. Material: blackheart malleable cast iron.

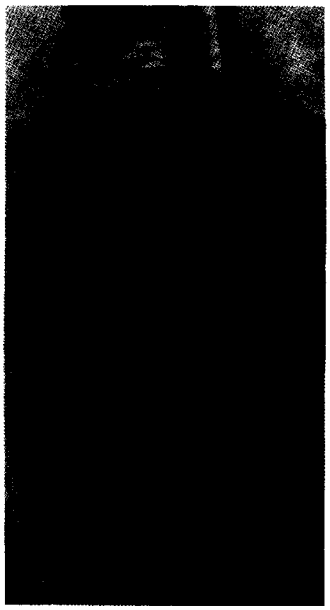
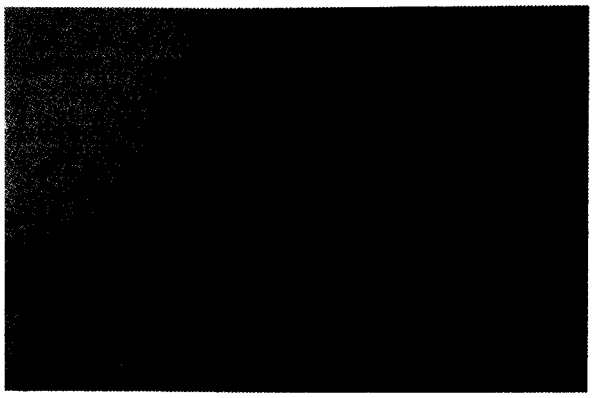


Fig. 28. Gear rack (material: P 45-06) to adjust the height of a plow. The tooth system is cast with high precision and machining is not required.



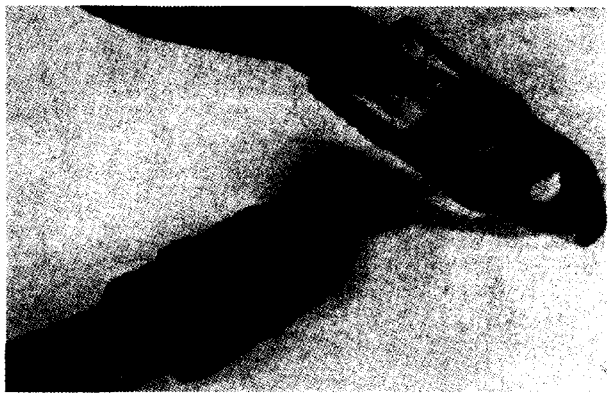


Fig. 29. Tractor brake lever bearing. Steel stamping and connection piece (B 35-10) welded on to a flat steel bar.

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the French Federal Railway (SNCF). A soft steel pipe with a 4-mm wall thickness is welded to the P 45-06 pearlitic malleable cast iron support. It is welded, without preheating, with a rutile-coated electrode.

The container foot in Fig. 27 comes from the transportation industry. This foot is welded to the sheet steel frame.

The examples of Figs. 28 and 29 are from the agricultural machine industry. Figure 28 shows a gear rack to adjust the height of a plow. This is a P 45-06 malleable iron casting and is welded to 10-mm plate steel. It is welded, without preheating, with a 4-mm diameter unalloyed basic-coated electrode. The tractor brake lever bearing in Fig. 29 consists of a steel stamping and a B 35-10 blackheart malleable cast iron connecting piece, both of which are welded onto a flat steel bar. This is also welded without preheating. The welding electrode contains 55-60% nickel.

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Electron Beam Welding of Pearlitic Malleable Iron

Fusion welding of malleable irons is a subject of interest to both producer and user but it presents four principal problems which limit successful applications.

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Abstract

This paper is a feasibility study of the joining of pearlitic malleable iron to pearlitic malleable iron as well as to three medium carbon steels by electron beam welding and analysis of resulting macro and microstructures. This method was found to offer a means of joining which minimizes fusion and heat treated zones as well as complicated joining preparation. However, direct joining of pearlitic malleable to pearlitic malleable by this process is unsatisfactory; joining of pearlitic malleable to plain carbon steels is not satisfactory either. Electron beam weldings of malleable to SAE 4140 and 4340 result in joints of normally acceptable soundness.

Introduction

FUSION WELDING OF MALLEABLE IRONS, a subject of interest to both producer and user, presents four principal problems which limit successful applications, namely 1) an extremely hard heat-affected zone, 2) a white iron structure in the fusion zone, 3) gas and slag generation during welding, and 4) inconsistency of results.

Conventional gas and coated or washed electrode manual arc welding with carbon steel filler rod results in rather large fusion and heat-affected zones. Weld soundness has been frequently less than satisfactory. However, for conditions of low stress, particularly in shock or fatigue and ones in which very occasional failures can be tolerated, the record shows many successful applications of manual arc welding.

Careful preparation and technique, use of low current, nickel-base welding rod, and rather extensive postheat treatment have been shown to produce welds in malleable iron with substantial mechanical properties.[1] Cary[2] has shown a series of successful techniques for the welding of malleable irons, cautioning that each application must be examined by its own merits and service requirements need to be evaluated on the basis of mechanical properties obtainable. Recent work by the Malleable Research and Development Foundation[3] has established satisfactory application of welding of pearlitic malleable iron to carbon steel in simulated automotive drive train components by careful choice of design of joint and utilization of automated MIG (inert gas, consumable metallic electrode) welding equipment. The use of low current and small diameter rod reduced the fusion and heat-affected areas. This, combined with proper joint design resulted in welds capable of withstanding a comparatively high level of cyclic stress. No postheat treatment was used.

Test Procedure

The feasibility study reported here is an outgrowth of a specific investigation of joining SAE 4340 steel to 70002 pearlitic malleable iron. The initial criteria of success, minimum heat affected zone, maximum retention of base material structure and/or properties and high level of weld soundness have been retained. The pairs of materials joined have been expanded to include 70002 pearlitic malleable to 1) 70002 pearlitic malleable, 2) SAE 1040 steel, and 3) SAE 4140 steel as well as the initial pair of materials. The only section thickness used for this work was $\frac{5}{8}$ in.

Electron beam welding was selected as the joining method because of the prospect of small heat-affected zone, absence of contamination from external sources, absence of need of a filler rod and simplicity of preparation of mating surfaces. The process has already proven itself in the joining of a wide variety of materials in a wide range of section thicknesses.

Stock preparation consisted of conventional quenching and tempering of steel stock with a final tempering temperature of 950 F (510 C) and quenching and

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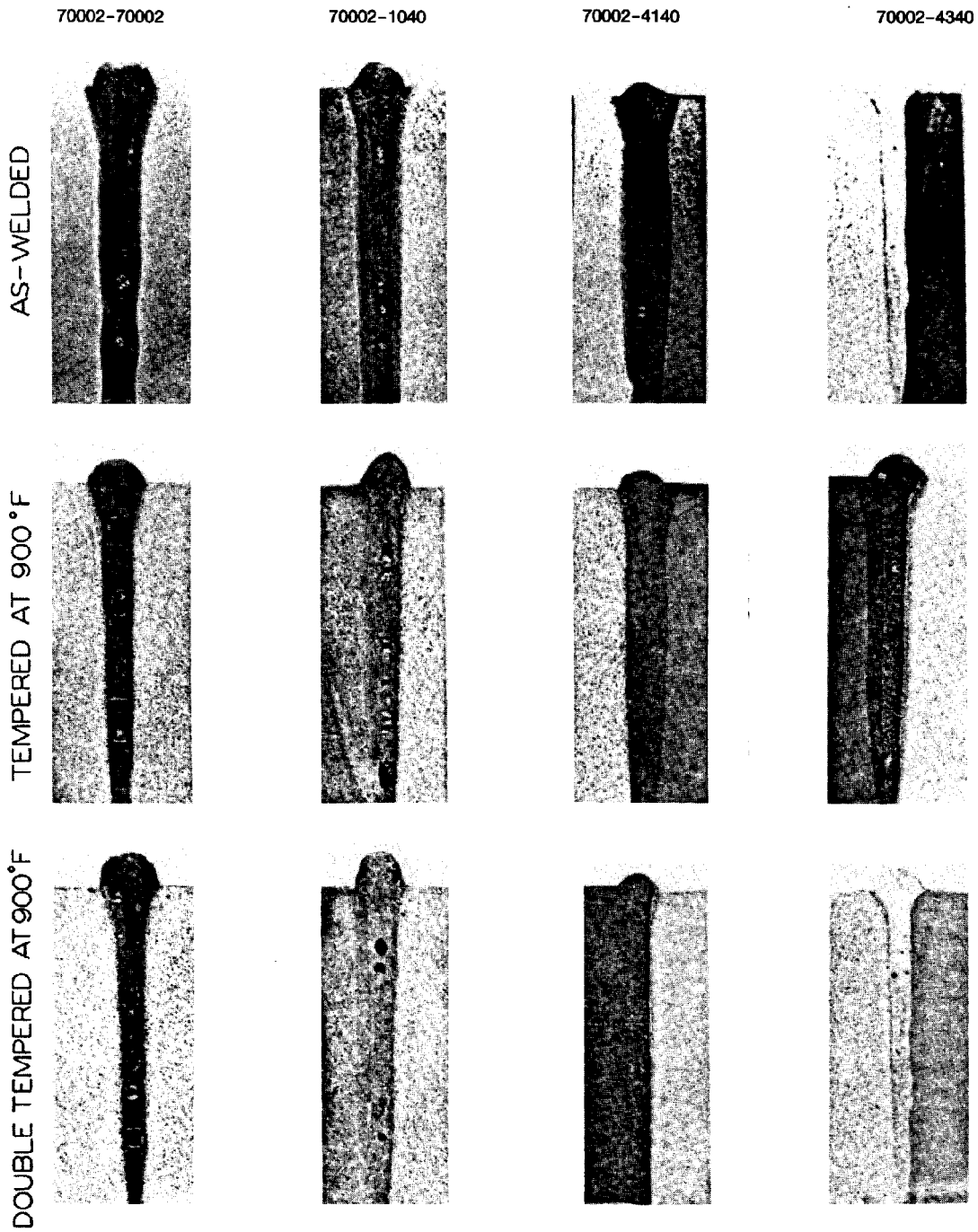


Fig. 1. Typical macro-etched cross-sections of electron beam welds. Actual thickness of cross-section is ⅝ in.

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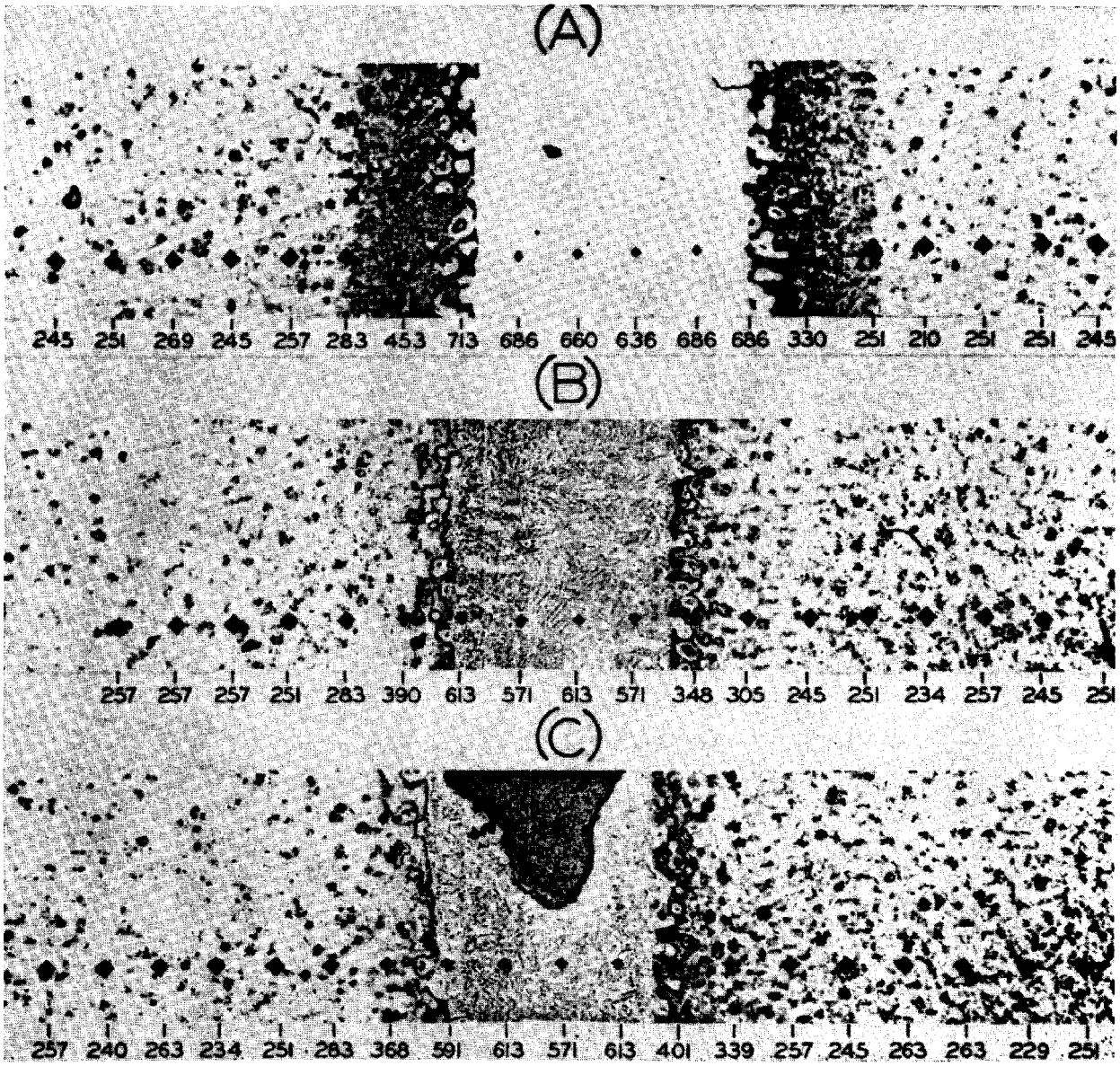


Fig. 2. Pearlitic malleable welded to pearlitic malleable; a) as-welded, b) tempered at 900 F (482 C), c) doubled at 900 F (482 C). Vickers diamond pyramid hardness values are shown at 0.010 in. intervals.

tempering a 2.50% carbon, 1.50% silicon, arrested anneal pearlitic malleable iron to a final hardness of 197-241 Bhn. Machine to ⅝ in. thickness, 1 in. width and approximately 4 in. length was then performed. Mating surfaces (⅝ x 1 in.) were finished by grinding.

The actual welding operations were conducted at Mech-Tronics Corp., Melrose Park, Ill. under the supervision of L. A. McNabb. The electron beam welding equipment used was equipped with a 50 x 30 x 42 in. vacuum chamber. Vacuum chamber pressure was maintained at 10^{-4}mm Hg for all welds. After initial experimentation, traverse rate was fixed at 40 in. per minute and heat input at 6700 watts. Parts were clamped in place with mating ends butted together for the welding operation.

One-third of the samples were tempered at 900 F (482 C) for one hour, one third were double tempered at 900 F (482 C) for one hour (each treatment) and the balance remained as welded. The double tempering treatment was instituted specifically for the 4340

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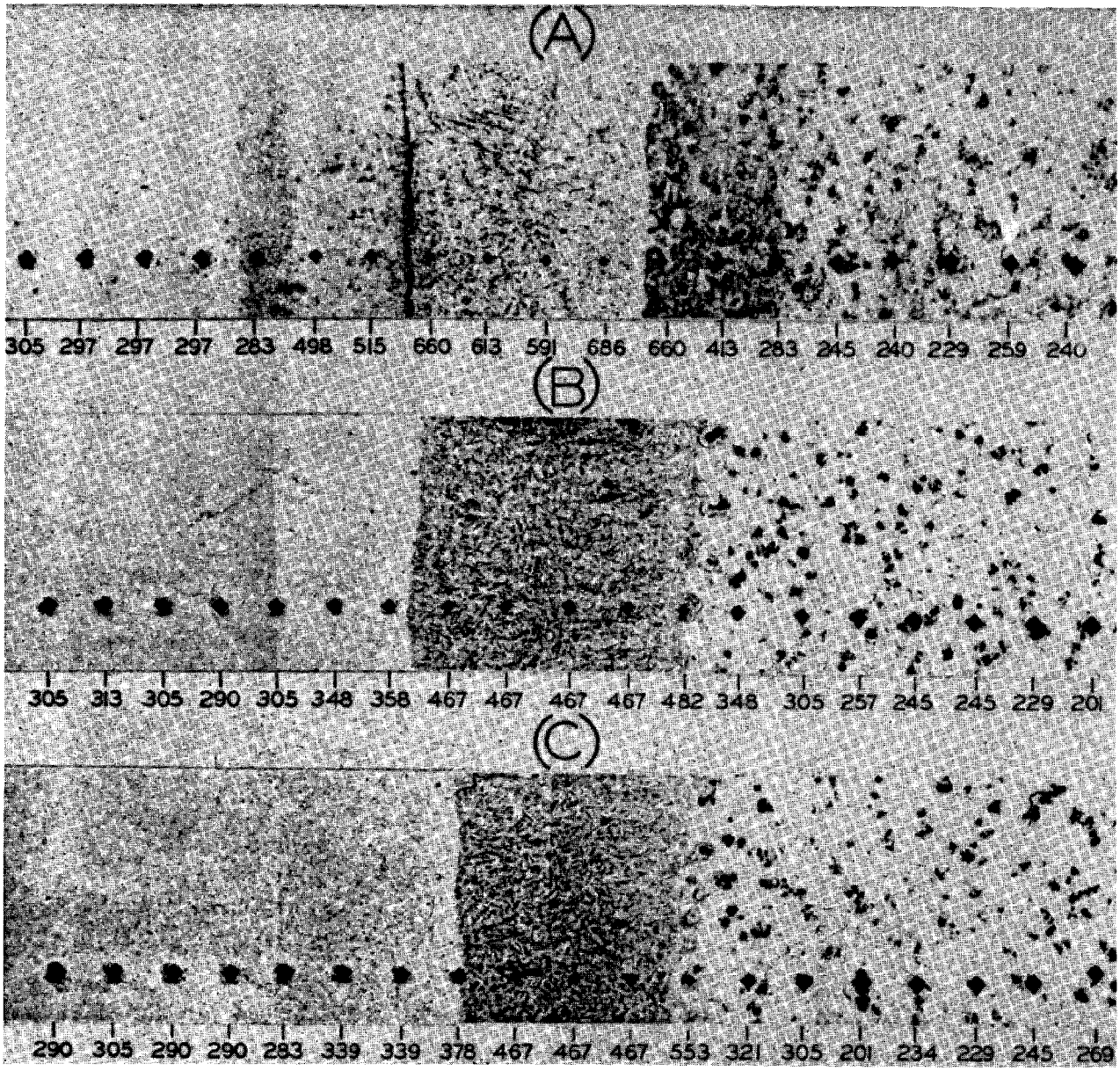


Fig. 3. Pearlitic malleable welded to SAE 1040 steel; a) as-welded, b) tempered at 900 F (482 C), c) double tempered at 900 F (482 C). Vickers diamond pyramid hardness values are shown at 0.010 in. intervals.

to 70002 weld because of the obvious retention of substantial austenite on the initial cooling and was included for other welds in the interest of providing consistent bases of comparison as well as evidence of any as-welded austenite retention. All samples were sectioned through the weld area and examined for weld soundness. A typical set of these samples were polished, etched, and subjected to a Vickers microhardness traverse (Vickers penetrator, 1 kilogram load).

Discussion of Results

Figure 1 illustrates typical cross-sections of the various pairs of mating materials and treatments. Welds between 70002 pearlitic malleable and 1) 70002 pearlitic malleable and 2) SAE 1040 steel were characteristically high in porosity while the joints between the pearlitic malleable and the two alloy steels were just as characteristically sound, with only an occasional sign of porosity. The strong deoxidizing effect of chromium and molybdenum in the

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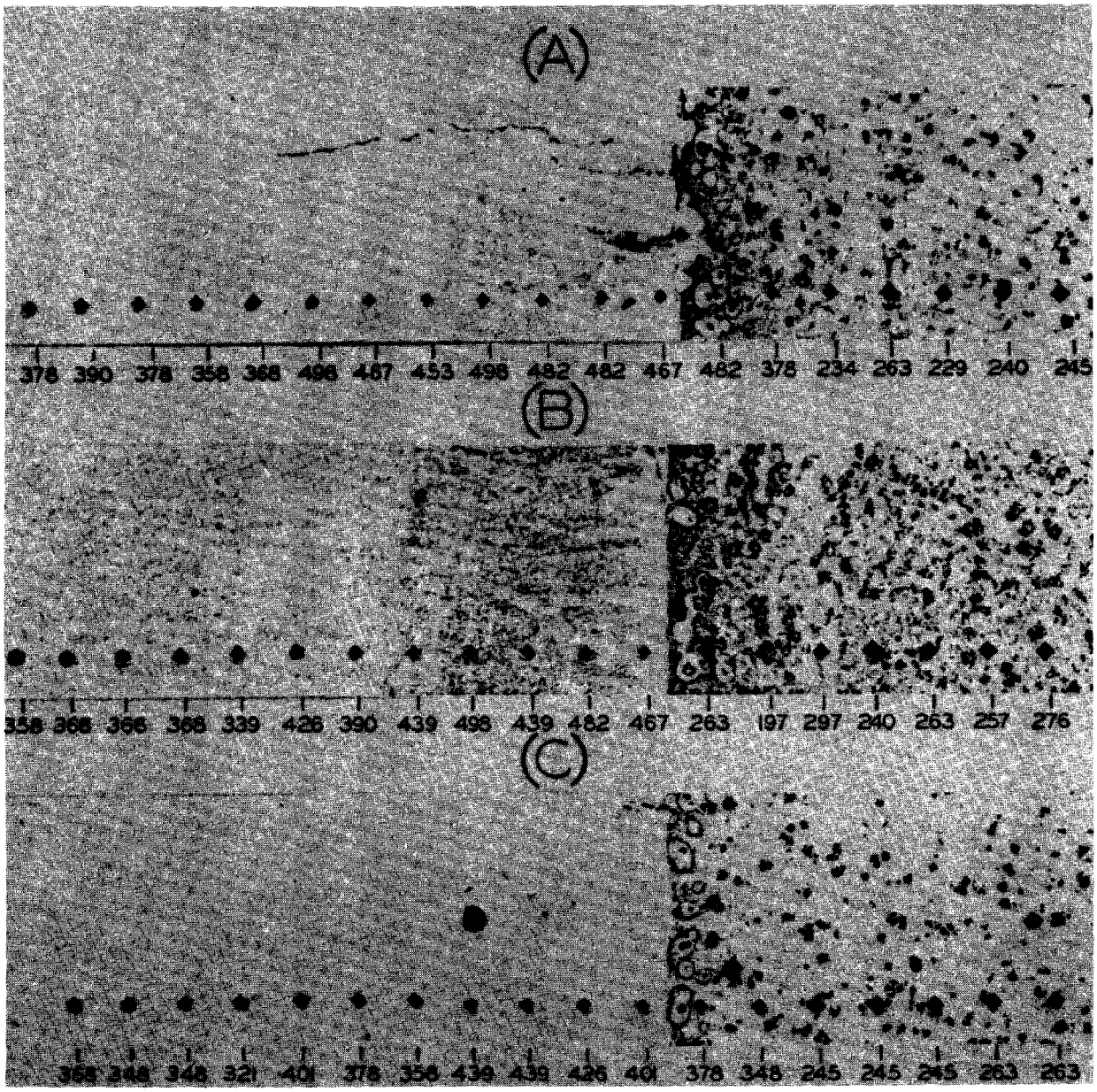


Fig. 4. Pearlitic malleable welded to SAE 4140 steel; a) as-welded, b) tempered at 900 F (482 C), c) double tempered at 900 F (482 C). Vickers diamond pyramid hardness values are shown at 0.010 in. intervals.

4140 and 4340 steels is credited with this difference. Characteristic of all welds was the very narrow (0.030 to 0.040 in.) fusion zone, a feature considered impossible with other fusion welding techniques.

Figures 2 and 5 depict the general microstructure and the Vickers hardness traverse values at 0.010 in. intervals across the weld and heat affected zones for all four pairs of materials with each of the three post-heat treatments. Magnification may be deduced from the spacing of the hardness indentations. Figures 6 and 7 more clearly show microstructures in significant areas of the joints and Fig. 8 is a graphical summary of hardness traverse values previously shown in Fig. 2 and 5.

In the joining of 70002 pearlitic malleable to itself (See Fig. 2) the fusion zone, both as-welded and after tempering, was characteristically white iron. The first tempering treatment substantially lowered the hardness of the white iron but the second temper produced no measurable effect, indicating absence of retention

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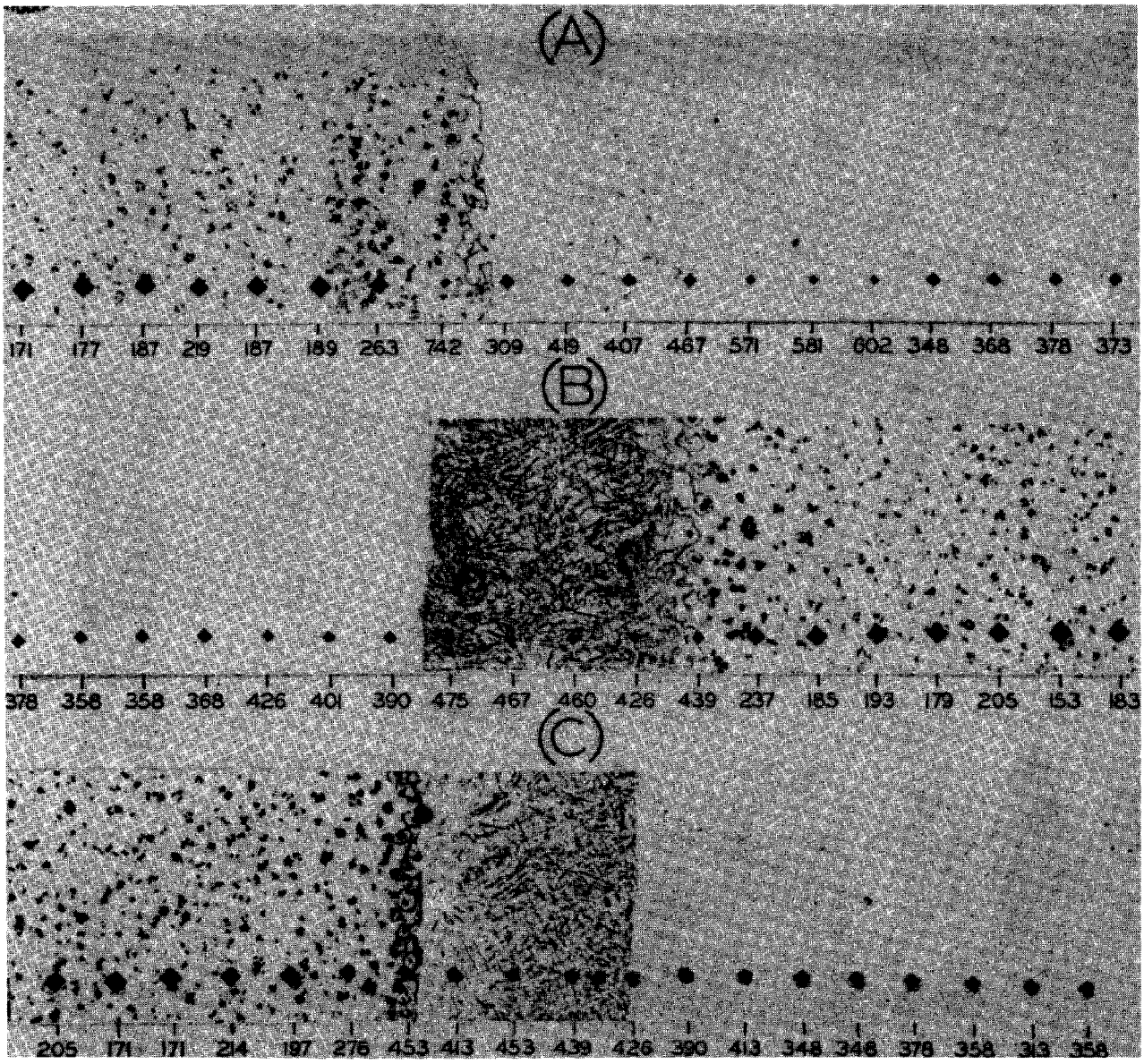


Fig. 5. Pearlitic malleable welded to SAE 4340 steel; a) as-welded, b) tempered at 900 F (482 C), c) double tempered at 900 F (482 C). Vickers diamond pyramid hardness values are shown at 0.010 in. intervals.

of austenite in the initial weld. Failure of the tempering treatments to reduce the hardness of the heat-affected zones to the level of base material hardness is ascribed to the low tempering temperature of the post-weld treatment as compared to the initial tempering temperature of the base material, 1300 F (704 C). However, no treatment short of re-annealing would have produced optimum microstructure in the weld zone. Significant microstructures, base material and heat-affected and weld zones in both as-welded and tempered condition are shown at higher magnification in Fig. 6 and 7.

Referring to Fig. 3 (pearlitic malleable welded to SAE 1040 steel) and the appropriate high magnification microstructures shown in Fig. 6 and 7, it may be seen that a) the as-welded fusion zone is martensitic with little or no austenite retention, b) the heat affected of the steel zone returns to base material hardness level upon tempering, c) the weld zone and the pearlitic malleable heat affected zone reach

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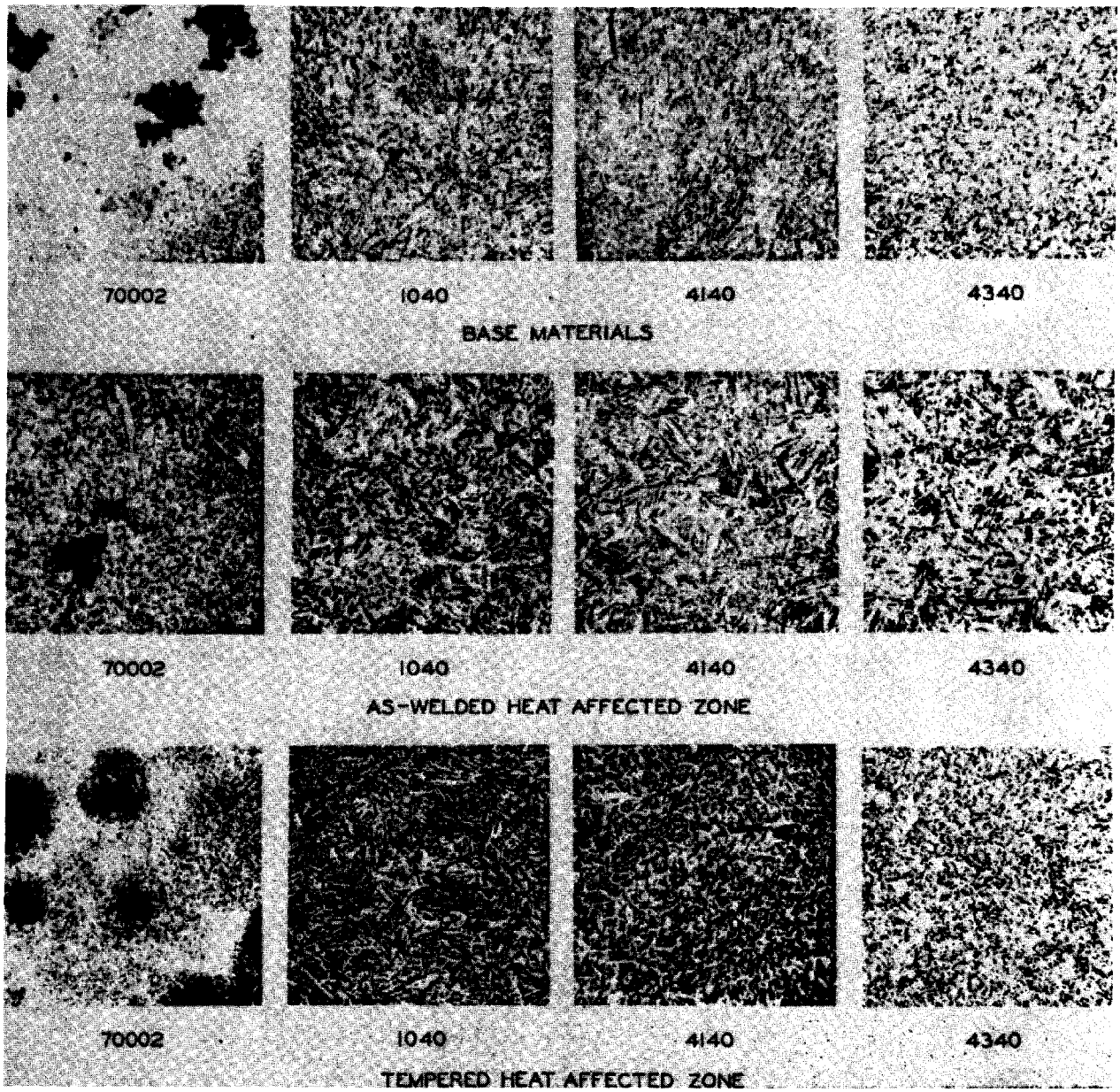


Fig. 6. Microstructures of base materials and heat affected zones. Original micros 3 x 3 in. Nital etch, 500×.

a machinable level of hardness after tempering at 900 F(482 C). It may be expected from the microstructures after the post-weld treatment that mechanical properties will be satisfactory for normal application. However, the basic defect of heavy porosity which was indicated earlier disqualifies the joint for such use.

Among the pairs of materials welded, the microstructure of the fusion zone of the 70002 to SAE 4140 steel is unique (See Fig. 4 and appropriate microstructures in Fig. 6 and 7. The principal constituent is upper bainite, rather than martensite. Some austenite retention is indicated by the drop in hardness after the second tempering operation. It might be expected that after a first tempering operation that the engineering properties of the joint might be satisfactory and that properties of strength and impact resistance of the weld and heat-affected zones should be at least comparable with the pearlitic malleable base material. The second tempering operation would further enhance this match of properties as well as improve machinability. The islands of white iron which appear in the heat-affected zone adjacent to the weld zone in the pearlitic malleable iron in this and the other welds are certainly detrimental in nature although to a much lesser degree than an entire weld zone of white iron. Further evaluation of this area is needed. The crack appearing in the as-welded structure is believed related to the degree of restraint during welding and, on admittedly meager

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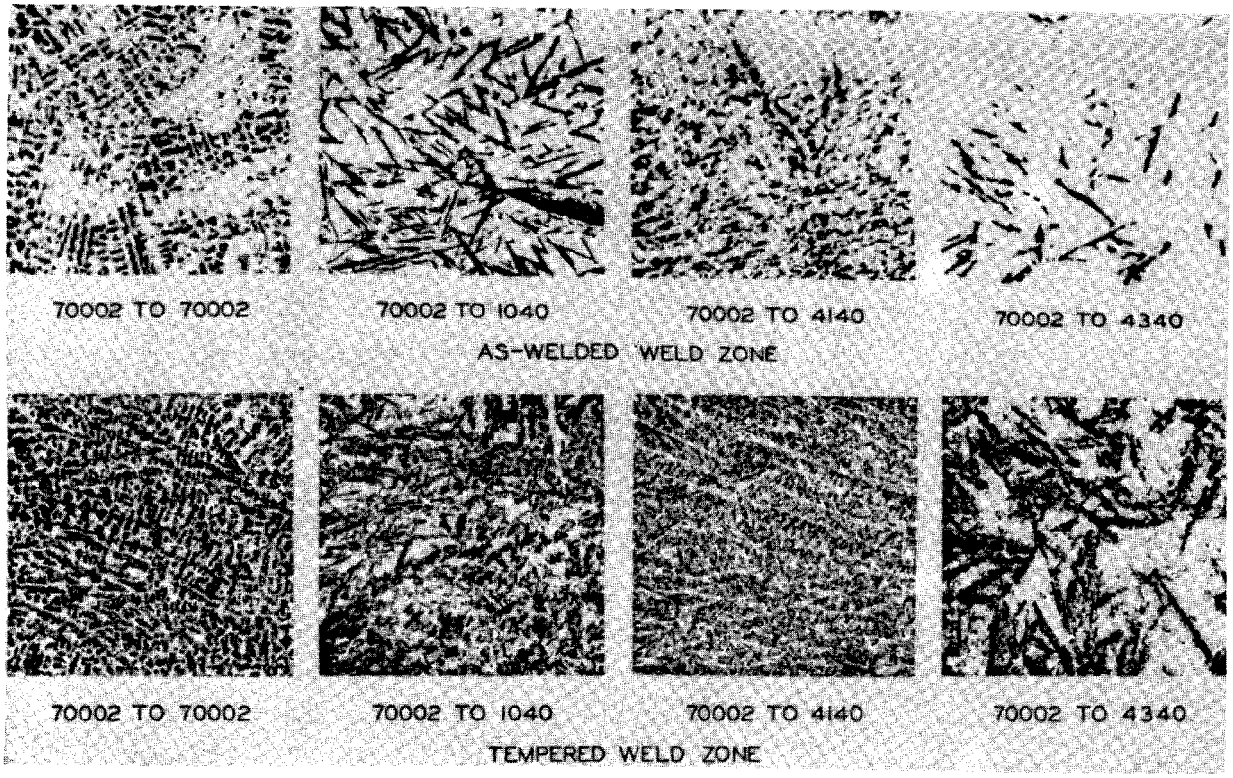


Fig. 7. Microstructures of as-welded and tempered weld zones. Original micros 3 x 3 in. Nital etch, 500×.

evidence, is not considered characteristic of the electron beam joining of these two materials.

Figure 5 and associated microstructures in Fig. 6 and 7 depict the results of joining 70002 pearlitic malleable to SAE 4340 steels. Increase in weld zone hardness after a first tempering operation is indicative of considerable austenite retention in the basically martensitic structure. Heat-affected zone responses are normal with substantial softening after the first tempering treatment and little if any change after the second. The combination of weld soundness and satisfactory structure and hardness after the second temper leads to the judgment that this joint is satisfactory for most engineering purposes. Modification of tempering temperatures may of course be made to suit individual circumstances.

It may be seen, therefore, that electron beam welding offers at least a partial solution to the problems of fusion welding of malleable irons to low alloy steels 1) extremely hard heat-affected zone, 2) white iron structure in the fusion zone, 3) gas and slag generation during welding, 4) inconsistency of results, 5) large fusion zone and 6) extensive post-weld treatment. A brief discussion of a proposed method of adaptation of the electron beam technique to malleable to malleable joints appears under the subsequent heading "Areas for Further Investigation!"

Figure 8 is a summary of hardness traverse values for the four joint combinations in the three treatment conditions, individual hardness readings having been given at the appropriate locations in Fig. 2 through 5.

Conclusions

1) Electron beam welding offers a method of joining which minimizes fusion and heat-affected zones as well as complicated joint preparation.

2) Direct joining of pearlitic malleable irons to pearlitic malleable irons by this process is unsatisfactory because of a) excessive weld porosity and b) the development of a fusion zone of white cast iron.

3) Joining of pearlitic malleable irons to plain carbon steels is unsatisfactory because of excessive weld porosity, although acceptable microconstituents are obtained.

4) Electron beam weldings of malleable irons to SAE 4140 and 4340 steels (probably because of presence of potent deoxidizers) results in joints of normally acceptable soundness and with microstructures which may be acceptable as-welded or with no more than post-weld tempering operations.

Areas of Further Investigation

Two areas of additional work are suggested. First, extension of the very limited indication of the deoxi-

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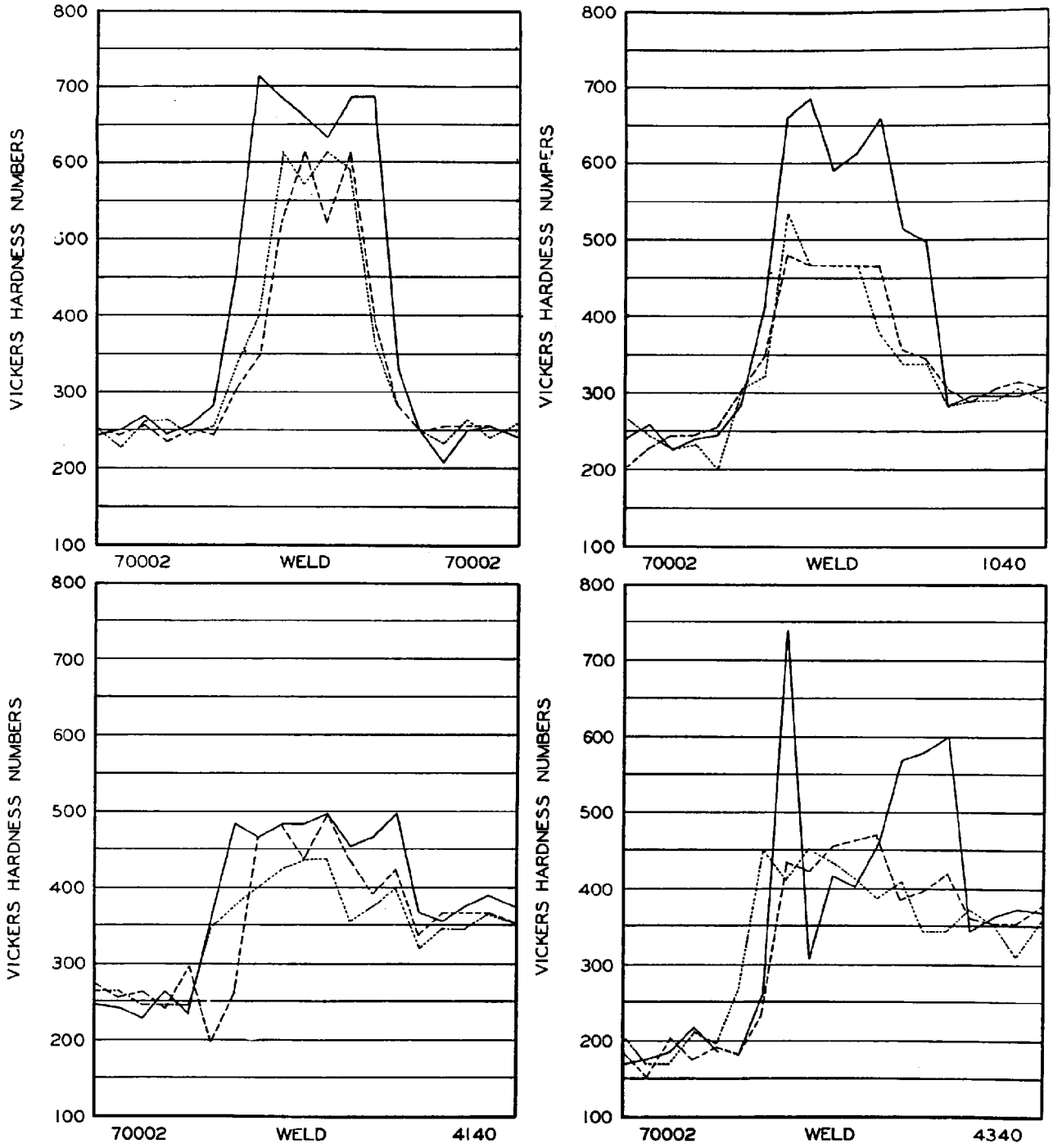


Fig. 8. Vickers diamond pyramid hardness vs. position across weld. For four material combinations. Legend: as-welded ———; tempered at 900 F - - - - - - - - - -; double tempered at 900 F — — — — — — —.

dation effect of the 4140 and 4340 steels should establish a most effective deoxidant for this type of joining. Second, additional work should certainly include the use of a sandwich technique, in which two pieces of malleable or pearlitic malleable iron are joined with an intervening layer of 0.015-0.020 in. of an appropriate compatible material such as a chromium-bearing steel. Fusion of base material to steel should be accomplished on both interfaces in a single weld pass. If this can be accomplished, a feasible means of welding malleable to malleable without serious degradation of mechanical properties may be achieved.

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Metal-arc welding of white cast iron

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Low heat-input technique allows successful repair of white cast-iron components

Problems involved in the metal-arc welding of martensitic white cast iron are discussed, together with the results of deposition tests intended to ascertain a suitable weld deposit length and electrode diameter. A newly developed weld deposit and build-up technique is described and the results of weld tests on white cast-iron test-blocks given.

Pearlitic grey cast iron, flame-hardened across the surface areas exposed to wear, may be the cheapest abrasion-resistant cast iron. At temperatures above 725°C, disaggregation of the pearlite results along with the transformation of ferrite to austenite, and the following water-quenching operation produces structures which are predominantly martensitic. As temperatures above 800°C do not usually occur during heating, and because the austenization period is short, dissolution of the free graphite is unlikely, and so the hardened structures probably contain little or no iron carbide.

Unalloyed white cast irons, when cast under suitable conditions, consist of a pearlitic matrix with a network of eutectic carbide. The greater hardness of the hyper-eutectic irons is due to the increased volume of eutectic carbide and the presence of primary cementite but, owing to the brittle condition of these irons, the hypo-eutectic compositions are usually employed. There is a number of white cast irons available to industry, some alloyed with nickel or chromium. The latter exhibit an outstanding resistance to abrasion due to a modification of the structure. The chromium addition replaces iron in the carbide phase, increasing the hardness to approximately 1200 DPN, while the nickel assists in producing martensitic structures by suppressing the austenite-to-pearlite transformation.

Tests have shown that, with regard to the problems of cracking during welding, there is little difference between alloyed and unalloyed white cast irons. Metallographic examinations indicate that the alloyed and unalloyed white cast irons are crack-sensitive even when a controlled heat input is maintained during welding. As would be expected, where high pre-heats are used (>450°C), the flame-hardened cast irons show a better response to welding because of a reduction in the hardness of the martensite due to its decomposition. However, high pre-heats on both alloyed and unalloyed white cast irons showed only a marginal reduction in the propensity to cracking, using conventional welding techniques.

Probable cause of cracking

Any non-ductile, crack-sensitive material must be considered as having a poor response to welding. It would appear that, during the welding of white cast iron, cracking is due to:

1) a localised heat input, which exceeds a threshold value (even with high pre-heats), causing expansion and contraction stresses that cannot be absorbed by the brittle casting;
2) the contracting weld bead creating high tensile strains on the heat-affected zone (HAZ) during cooling.

Obviously, hydrogen-induced HAZ cracking must be considered, although this effect can be avoided.

Low heat-input welding technique

An experimental weld-repair test was carried out on an alloyed white cast iron measuring 9 × 6 × 3in (229 × 152 × 76 mm). A defect was cast in the centre (figure 1), this measuring 3 in (76 mm) diameter × $\frac{5}{8}$ in (15.9 mm) deep. This experiment was prompted by some promising repairs achieved on martensitic white cast-iron components in the motor industry.

Two metal-arc electrodes were specially produced for these tests, and designated as:

0422—nickel based, containing a controlled percentage of iron; and

422—alloyed high nickel type.

Both electrode coatings were formulated for low heat input: 0422 was used for the initial layer, and a weld deposit length of $\frac{3}{4}$ in (19 mm) was adhered to strictly. The electrode diameter in all cases was 3.2 mm.

The first deposit (figure 2), was peened using a 2lb (0.91 kg) ball-pane hammer, with simultaneous air quenching. The air quenching time was 4 - 6 seconds, so the actual repair progressed rather quickly (figure 3), until the entire area including the outer edge, was clad (figure 4). The overall casting temperature did not rise above 40°C at any time. At this stage of the simulated repair, the base metal was protected from direct contact with the arc by a 3.5 mm layer of weld-metal. However, greater deposit lengths were not used because of increased heat-input cracking dangers.

The use of a single alloy only was unsatisfactory for a large repair, and so I developed a cross technique, which involved depositing a "cross" using 422 (figure 5), and filling in the intercross areas with 0422 (figure 6). Further crosses were built up at different positions to the underlying crosses, to avoid having a high nickel layer right through the repair.

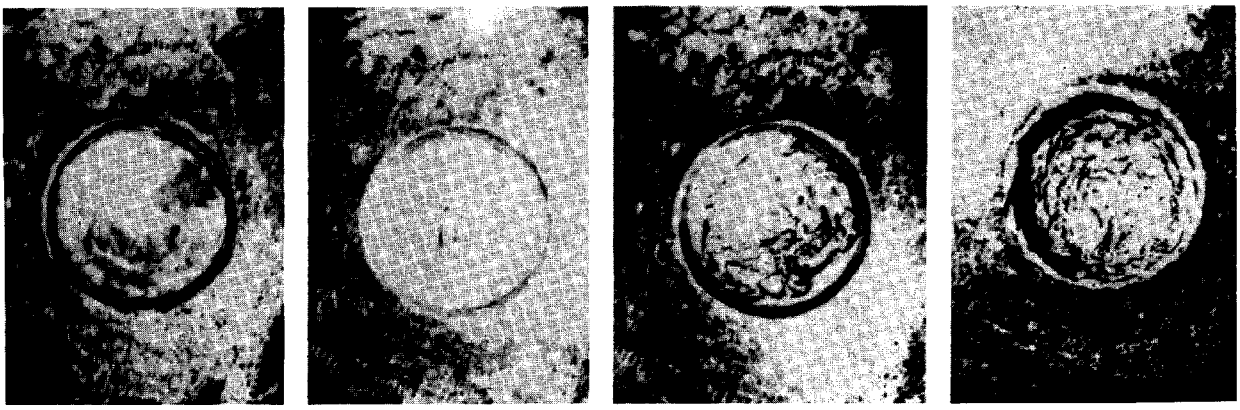
Because the surface weld-metal must exhibit a high resistance to abrasion, a chromium-carbide deposit was used for the two surface capping layers (figure 7), again employing the short deposit technique with air quenching. However, this technique promoted fine surface cracks, although present tests indicate that the effect can be reduced by using a wear-facing alloy of a different chemistry.

Results of experimental weld repair

The repair on the block of alloyed white cast iron was examined metallurgically. Surface "bleed-outs" were shown by dye-penetrant testing (as expected with the air quenching or

Source: Metal Construction and British Welding Journal, October 1973 (The Welding Institute)

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Figs. 1 to 4 (left to right). Fig. 1—Defect cast in white cast iron specimen. Fig. 2—The first deposit was peened. Figs. 3 and 4—The repair progressed quickly.

cold welding technique with a chromium carbide deposit). Radiological examination, using gamma rays, indicated the possibility of three cracks, $\frac{1}{4}$-$\frac{5}{8}$ in (6.4 - 15.9 mm) long, though this could not be confirmed. Figure 8 shows a section through the block. The small cracks at the weld/base junction could be due to welding, and not to the susceptibility of the hard irons to scorch cracking, but this could not be confirmed either, because of problems associated with cutting through a large block of hard iron.

The section (figure 8) reveals etchant-seepage from apparent cracks at the interface. Although the overall result was extremely promising, doubts as to whether or not the cracks were caused by welding, led to further investigation.

Cracking tests

To establish the cause of possible toe-cracking, extensive bead-on-plate tests were conducted using different weld-deposit lengths, with both 2.4 mm and 3.2 mm diameter electrodes. Cracking occurred in all cases, even with curtailed deposits (figure 9). Because of difficulties in sectioning the test runs, the only mode of cracking investigated initially was "toe-cracking", i.e., removal of the surface weld bead followed by dye-penetrant testing.

After similar tests, during which peening was carried out, cracking did not occur (figure 9). This indicated that cracking was due to contraction of the weld bead during structural transformations in the HAZ, or shortly afterwards when the contraction in the weld bead reached a peak. To confirm this, weld deposits ranging from 60 mm in length were made and peened immediately after the welding operation. Mixed results were obtained, probably due to the difficulties of peening sufficiently close to the weld pool (i.e., within 0.3 in, 7.6 mm) under these conditions. The results indicated that cracking was prevented where adequate peening had been managed, while severe cracking occurred where the weld bead had been missed; slight cracking resulted from insufficient peening.

Though this technique can be investigated more scientifically, it would appear that cracking may be caused by the contracting weld bead creating strains in the HAZ during a phase change (probably the $\gamma \rightarrow$ martensite + FeC_3). These strains were mainly eliminated by peening, provided the time lapse was not exceeded. Tests showed that in order to achieve a sound weld deposit, the critical time lapse before peening was seven seconds.

Although this technique may seem fraught with practical difficulties, it was, in fact, readily workable; a weld pad 1 in (25.4 mm) square was deposited in five minutes with a 2.4 mm diameter electrode. The cracking problems apply to the clad layer only, after which peening is used merely to reduce the contractional stress in the weld-metal on large fills, and at this stage the electrode diameter and deposit-length can be increased.

Cracking due to high differential heat input did not occur, because:

a) each test was conducted on a large section of alloyed white cast iron;
b) short deposits were made; and
c) air quenching was applied.

Quenching of the weld bead after each deposit has proved conclusively to overcome this danger of cracking.

A section through a weld bead was seen to be free of underbead cracking, with excellent fusion to the cast iron. The welding parameters in this instance were:

2.4 mm electrode diameter,
travel rate—6 in/min (152 mm/min),

Figs. 5 to 7 (left to right). A "cross" technique was developed.



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1666 Joules/in (65.6 Joules/mm),
deposit length 0.4 in (10.1 mm),
7s time lapse prior to peening.

The test results show a high susceptibility to cracking when using a 3.2 mm diameter electrode, unless a pre-layer is first achieved with a 2.4 mm diameter electrode. A 2.4 mm diameter electrode gave consistently good results, with deposit lengths of 0.4 in (10.1 mm) and a delay prior to peening of 7s. The travel rate during the tests (figure 9) was constant at 6 in/min (152 mm/min), and the heat input was calculated at 1666 Joules/in (65.6 Joules/mm) and 3000 Joules/in (118 Joules/mm), using 2.4 mm and 3.2 mm diameter electrodes respectively.

Substantial peening of a nickel-base weld bead is considered by some to be detrimental. However, extensive tests have shown this opinion to be unfounded. The weld deposit studied in section, was extensively peened, and figure 10 reveals an area from the weld surface.

Hardness surveys were carried out using a 0.5 kg load on a Zwick low-load hardness tester. The results are given in Table 1.

Table 1

Position	Hardness values DPN		
	Weld-metal	H.A.Z.	Parent metal
A	378 · 365 214 · 350	558 · 838 715	586 · 610 · 513
B	241 · 195 301 · 402 526 · 434	770 · 458	552 · 438 · 400
C	350 · 371 318 · 386	545 · 790 746	508 · 532 552 · 473

Figure 10 shows the structure, which is a nickel solid solution (422) with intergranular carbide.

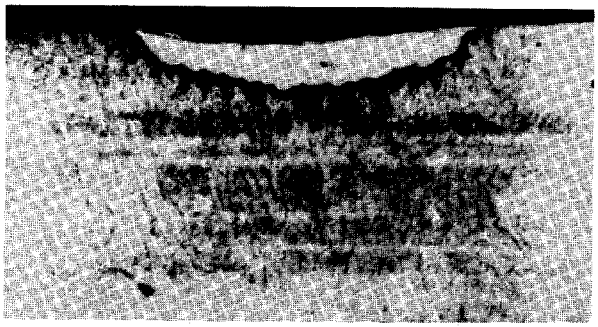
Repair and reclamation of castings

The metal-arc welding of white cast iron could probably take three forms:

1) repair of casting defects;
2) repair of damaged components in service; and
3) the rebuilding of worn surfaces.

The repair of casting defects would only apply when the defect was in a non-critical area of wear, because of the need for a substantial layer of relatively soft nickel-based weld-metal. A number of hard capping layers can be made to provide a wear resistance equal to the base metal, but the need for a soft clad layer (ratio of hard : soft = 3:1), does reduce the overall wear efficiency. This obviously applies more to the repair of worn surfaces and, in my opinion, the repair of white cast iron in this respect must be considered unsuitable. The repair of a damaged component in an area subject to low wear rates, or where the three-layer capping deposit will suffice, is a viable proposition, and there is a great deal of scope in this.

Fig. 8. Section through test block.



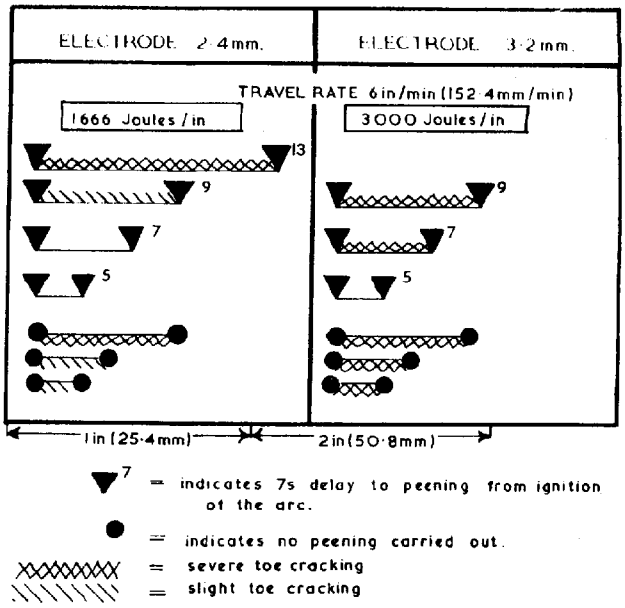


Fig. 9. Peening helped obviate cracking.

The choice of a capping electrode depends on the need to obtain a deposit having the minimum of fissures. Where surfaces fissures are not too important (the nickel-clad layer has been shown to halt their propagation), a deposit of high hardness may be used.

Practical weld test

In view of the visually promising results obtained in the tests, a breaker bar in a rock-crushing machine was repaired. Figure 11 shows the four defects, the two in the centre being adjacent. The defect faces were first clad using 422 and the technique previously described. The two smaller defects were clad with 422 followed by 0422 (figure 12), to leave sufficient for two layers on the wear-facing alloy. The completed repair is shown in figure 13.

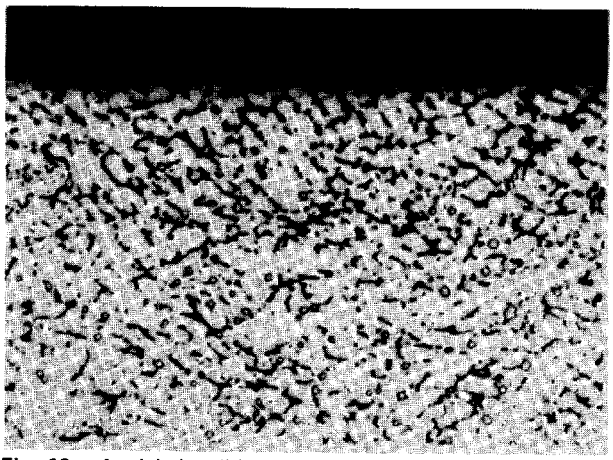


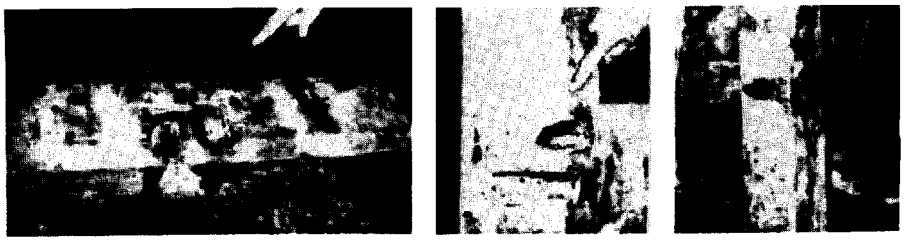
Fig. 10. A nickel solid solution with intergranular carbide.

Repair of cracks

The repair of cracks in alloyed white cast-iron components is more difficult than filling a defect but, at the same time,

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Figs. 11 to 13. Four defects clad successfully.



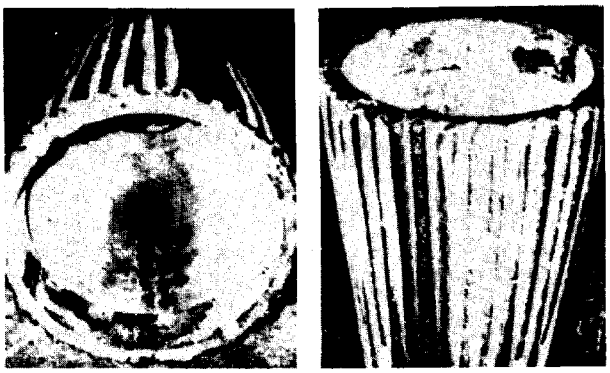
the crack may be in an area not subject to wear, and so this offers a viable and economical repair proposition.

The crusher roll in figure 14 had failed in service, a crack extending a considerable way down its length. The crack was carefully ground out, using a high-speed grinder and tungsten carbide tools, with the tied-in section "key-holed" to avoid a notch-sensitive area. In view of the difficulties involved, i.e., the thickness of the section, a modified deposit technique was employed.

A 2.4 mm diameter 422 was used at a low current, i.e., d.c. positive 30A. The deposit so produced provides islands of weld-metal, the current being too low for a continuous deposit.

The current was increased to 60A after the entire joint had been covered in this way, using the short-deposit quench technique. The second layer melted partly through the first and filled in the vacant areas, again at a low heat input to the base metal.

The joint was completed using a 3.2 mm diameter 0422 and 3.2 mm diameter 422 in alternate layers, because the repair was not suitable for the "cross" technique. The completed repair is shown in figure 15 which also illustrates a weld build-up on a crushing rib, using a clad layer of 422 followed by a two-layer build-up of a chromium carbide deposit. In all cases, the short deposit and peen technique was used. It should be noted that, as stated earlier, the economics of extensive overlaying, such as the hard deposits on the crushing rib, are doubtful.



Figs. 14 and 15. Repair of crusher roll.

Conclusions

The tests showed that the metal-arc welding of white cast iron can be carried out with the promise of success. The optimum weld-deposit length for the production of a deposit free from base-metal cracking was established, and numerous successful repairs carried out.

Acknowledgements

Thanks are due to Mr. O. P. Jain (metallurgist), and to the Eutectic | Castolin Institute for the production of the special electrodes used in this work.

Reference

1. Cookson, C.: "The metal-arc repair welding of cast iron", presented to the Irish Welding Association, January, 1971.

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Weldability of Spheroidal-graphite-type Ni-Resist Cast Irons*

By S. N. Anant Narayan, DSc., A. J. Rickard, MSc., FIM, MIBF. and N. Stephenson, BMet., PhD

Austenitic cast irons with either flake or spheroidal graphite, and frequently known under the generic title of Ni-Resist, are finding increasing application for corrosion resistant applications, particularly for pumps and valves and sealing rings required in marine and brackish water environments, the production of potable water from the sea, and for the provision of cooling water for the power generating, petroleum and chemical industries. Additionally, these alloys have good resistance to heat and oxidation which makes them suitable for turbo-charger casings, diesel engine manifolds and heat-treatment equipment. Occasionally, it is necessary for founders and end-users to weld these irons to rectify casting defects and machining errors. Also, some fabrication may be necessary to enable lugs etc. to be attached. The flake graphite grades can be welded with little difficulty using 55% nickel/iron/carbon MMA coated electrodes given adequate pre-heat and interpass temperatures, provided the sulphur and phosphorus contents are each controlled to 0.04% maximum. Austenitic irons having a spheroidal-graphite structure have been observed to be less weldable and this article describes how the problem has been largely overcome in those grades containing chromium through compositional control, the addition of niobium and the use of selected welding consumables of the 55% Ni/Fe/C type. It has been shown, furthermore, that pre-heating is unnecessary and that fabrication is potentially practicable.

cast irons irrespective of whether the graphite be in the flake or spheroidal form. These irons, described in detail elsewhere,[1] are being used increasingly in the marine, desalination, power, petroleum and chemical industries, especially for casings of pumps and valves and for sealing rings, by virtue of their corrosion resistance, particularly in sea and brackish waters. They also have good resistance to heat and oxidation which makes them suitable for turbo-charger casings, diesel engine manifolds and heat treatment equipment. Furthermore, by appropriate selection of composition, the coefficient of thermal expansion can be varied over a wide range making them compatible for service in conjunction with other materials such as aluminium alloys and steels.

Like unalloyed cast irons, most of the Ni-Resist irons can be produced with either flake or spheroidal Ni-Resist is the name applied to a family of alloyed cast irons in which the presence of between 14 and 36% nickel in conjunction with, in some instances, copper at the lower nickel contents, has produced a fully austenitic structure and conferred mechanical and physical properties unobtainable from ordinary

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* This paper was originally presented to the Indian Institute of Welding, November 1977, and to the Afro-Asian Conference on Welding Metals Technology, India, February 1978. Of the authors, Dr. Anant Narayan is attached to International Nickel Services (UK), Limited, and Mr. Rickard and Dr. Stephenson to Inco Europe, Limited.

Source: Fountry Trade Journal, July 5, 1979

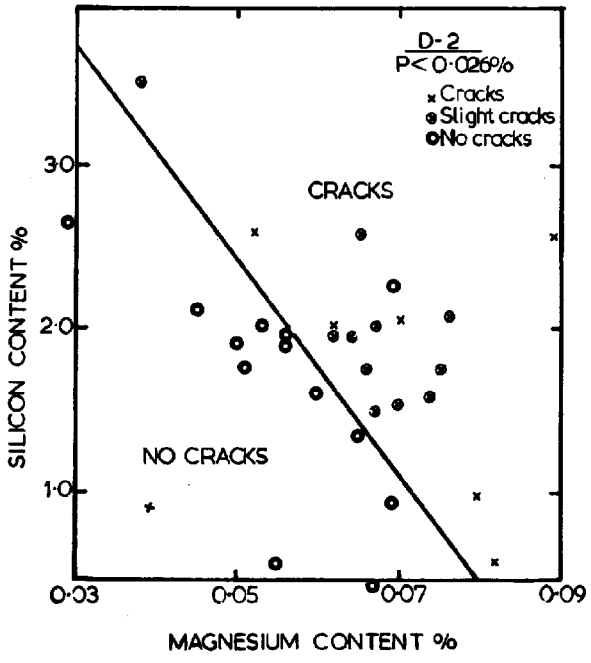
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TABLE 1: *ISO and Related National Specifications of Ni-Resist Austenitic Cast Irons.*

Grade	C max	Si	Mn	Ni	Cr	Cu	P max	0·2% proof stress (Rp0·2) N/mm²	T.S. (Rm) N/mm²	Elong. (A) 5·65 √So %	Mean impact value: Charpy V-notch J	Mean impact value: U-notch (Mesnager) J
ISO 2892: 1973 (E)												
Flake graphite Ni-Resists												
	Composition per cent							Mechanical properties (minimum)				
L-NiCuCr 15 6 2	3·0	1·0-2·8	0·5-1·5	13·5-17·5	1·0-2·5	5·5-7·5	—	—	170	—	—	—
L-NiCuCr 15 6 3	3·0	1·0-2·8	0·5-1·5	13·5-17·5	2·5-3·5	5·5-7·5	—	—	190	—	—	—
L-NiCr 20 2	3·0	1·0-2·8	0·5-1·5	18·0-22·0	1·0-2·5	0·5 max	—	—	170	—	—	—
L-NiCr 20 3	3·0	1·0-2·8	0·5-1·5	18·0-22·0	2·5-3·5	0·5 max	—	—	190	—	—	—
L-NiCr 30 3	2·5	1·0-2·0	0·5-1·5	28·0-32·0	2·5-3·5	0·5 max	—	—	190	—	—	—
L-NiSiCr 30 5 5	2·5	5·0-6·0	0·5-1·5	29·0-32·0	4·5-5·5	0·5 max	—	—	170	—	—	—
L-Ni 35	2·4	1·0-2·0	0·5-1·5	34·0-36·0	0·2 max	0·5 max	—	—	120	—	—	—
Spheroidal graphite (ductile) Ni-Resists												
S-NiCr 20 2	3·0	1·5-3·0	0·5-1·5	18·0-22·0	1·0-2·5	0·5-max	0·08	210	370	7	13	16
S-NiCr 20 3	3·0	1·5-3·0	0·5-1·5	18·0-22·0	2·5-3·5	0·5 max	0·08	210	390	7	—	—
S-Ni 22	3·0	1·0-3·0	1·5-2·5	21·0-24·0	0·5 max	0·5 max	0·08	170	370	20	20	24
S-NiMn 23 4	2·6	1·5-2·5	4·0-4·5	22·0-24·0	0·2 max	0·5 max	0·08	210	440	25	24	28
S-NiCr 30 1	2·6	1·5-3·0	0·5-1·5	28·0-32·0	1·0-1·5	0·5 max	0·08	210	370	13	—	—
S-NiCr 30 3	2·6	1·5-3·0	0·5-1·5	28·0-32·0	2·5-3·5	0·5 max	0·08	210	370	7	—	—
S-NiSiCr 30 5 5	2·6	5·0-6·0	0·5-1·5	28·0-32·0	4·5-5·5	0·5 max	0·08	240	390	—	—	—
S-Ni 35	2·4	1·5-3·0	0·5-1·5	34·0-36·0	0·2 max	0·5 max	0·08	210	370	20	—	—
S-NiCr 35 3	2·4	1·5-3·0	0·5-1·5	34·0-36·0	2·0-3·0	0·5 max	0·08	210	370	7	—	—

The BSI specification requirements are the same as those in the ISO 2892 Standard.
DIN 1694 also includes a weldable grade viz. GGG-NiCrNb 20 2. Werkstoff Nr. 0·7659.

1 Effects of silicon and magnesium on HAZ cracking in groove welds made in low-phosphorus, niobium-free iron D-2.



Weldability of Spheroidal-graphite-type Ni-Resist Cast Irons

graphite, and the latter types have tensile strengths equal to about twice that of the flake graphite equivalents. However, spheroidal graphite cannot be produced in the 6-7% copper-bearing grades. Compositions and properties of the several grades of Ni-Resist iron together with details of the related national and international specifications appear in Table 1.

Flux-coated, manual metal-arc (MMA) electrodes of the 55% nickel/iron/carbon type, widely used for the reclamation and repair-welding of unalloyed grey and s.-g. iron castings are also employed for similar operations on Ni-Resist irons if the presence of ferro-magnetic weld metal can be accepted. Mild steel electrodes cannot be used because, as with ordinary grey irons, a brittle weld will result from carbon pick-up. Moreover, the weld would have inferior resistance to corrosion and heat. Oxyacetylene welding using a filler metal similar in composition to the base metal can be undertaken but is not strongly recommended because of difficulties experienced in controlling casting temperatures and cooling rates and the risk of cracking. The 55% nickel/

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TABLE 1: (*continued*).

Grade	Related specifications				
	USA	UK	France	Germany	
	ASTM A436-72 Type	BSI 3468(1974)[1] Grade	AFNOR NF-A 32-301 (1972) Grade	DIN 1694 Grade	DIN 17007 Werkst. f. Nr
L-NiCuCr 15 6 2	1	L-NiCuCr 15 6 2	L-NUC 15 6 2	GGL-NiCuCr 15 6 2	0·6655
L-NiCuCr 15 6 3	1b	L-NiCuCr 15 6 3	L-NUC 15 6 3	GGL-NiCuCr 15 6 3	0·6656
L-NiCr 20 2	2	L-NiCuCr 20 2	L-NC 20 2	GGL-NiCr 20 2	0·6660
L-NiCr 20 3	2b	L-NiCr 20 3	L-NC 20 3	GGL-NiCr 20 3	0·6661
L-NiCr 30 3	3	L-NiCr 30 3	L-NC 30 3	GGL-NiCr 30 3	0·6676
L-NiSiCr 30 5 5	4	L-NiSiCr 30 5 5	L-NSC 30 5 5	GGL-NiSiCr 30 5 5	0·6680
L-Ni 35	5	L-Ni 35	L-N 35	GGL-Ni 35	0·6683
	ASTM A 439-71 Type				
S-NiCr 20 2	D-2	S-NiCr 20 2	S-NC 20 2	GGG-NiCr 20 2	0·7660
S-NiCr 20 3	D-2B	S-NiCr 20 3	S-NC 20 3	GGG-NiCr 20 3	0·7661
S-Ni 22	D-2C	S-Ni 22	S-N 22	GGG-Ni 22	0·7670
Si-NiMn 23	(ASTM A 571-71 Type D-2M)	S-NiMn 23 4	S-NM 23 4	GGG-NiMn 23 4	0·7673
S-NiCr 30 1	D-3A	S-NiCr 30 1	S-NC 30 1	GGG-NiCr 30 1	0·7677
S-Ni 30 3	D-3	S-NiCr 30 3	S-NC 30 3	GGG-NiCr 30 3	0·7676
S-NiSiCr 30 5 5	D-4	S-NiSiCr 30 5 5	S-NSC 30 5 5	GGG-Ni Si Cr 30 5 5	0·7680
S-Ni 35	D-5	S-Ni 35	S-N 35	GGG-Ni 35	0·7683
S-NiCr 35 3	D-5B	S-NiCr 35 3	S-NC 35 3	GGG-NiCr 35 3	0·7685

iron/carbon MMA electrode can be used also for joining the irons themselves, to mild steel, low alloy constructional steels, some stainless-steels and to simple high-nickel alloys.

Given adequate pre-heat (300-350°C) and interpass temperature in the same range, few problems have arisen during simple welding operations on the flake graphite grades of Ni-Resist iron which are tougher and more ductile than ordinary grey cast irons. Furthermore, despite their higher expansion coefficients and lower thermal conductivity, the Ni-Resist irons have proved to be more tolerant to stresses induced by expansion and contraction during welding. The austenitic matrix implies less tolerance for impurity elements such as phosphorus and sulphur and to avoid hot-cracking it is strongly recommended that the amounts of each of these elements present in flake graphite grades of Ni-Resist cast iron should not exceed 0.04%, preferably less than 0.03%. Such limitations restrict the melting of such irons to furnaces other than the cupola unless desulphurisation techniques are used.

The greatly enhanced ductility and toughness of the s.-g. grades of Ni-Resist might suggest weldability superior to that of the flake graphite grades. Experience has shown that this is not necessarily so, for castings of certain types, notably D-2, D-2C and D-2M, have displayed cracks up to 3mm. long in the weld heat-affected zone and normal to the fusion boundary.

The effects of compositional variations and welding procedures on the incidence of the defect, investigated by Inco, led in recent years to the finding that the cracking was greatly diminished or eliminated by making small additions of niobium to castings of the above three types. Further work has now been undertaken to define the compositional ranges associated with improved weldability of s.-g. Ni-Resist irons. Additionally, in relation to the desirable extension of welding operations to heavy sections, to fabrication, and to automatic processes, all conceivably associated with higher design stresses, a number of welding consumables and techniques have been assessed and the mechanical properties of weldments evaluated.

Effects of Compositional Variations

The effect of composition on weldability had been earlier investigated in Inco laboratories in X-welds made between cast 25mm. square bars of s.-g. Ni-Resist iron placed edge-to-edge. Transverse sections were strained to open up any HAZ cracks present to enable the quality of the weld to be assessed by examination under a lower-power microscope (×15),

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TABLE 2: *All-weld-metal compositions.*

Consumable identity	Diameter mm	All-weld-metal composition %								
		C	Mn	Si	S	P	Ca	Mg	Co	Cr
A	3·25	1·1	0·39	0·72	0·001	—	0·02/0·03	<0·01	0·12	—
	4·0									
B	3·25	1·3	0·44	0·94	0·001	—	c. 0·005	<0·01	<0·05	—
D	3·25	1·03	0·31	<0·05	0·002	0·006	c. 0·003	0·004	0·07	<0·01
E	3·25	0·88	0·71	0·68	0·002	0·007	c. 0·002	0·015	0·12	0·11
F	4·0	1·35	0·41	0·54	0·002	0·005	0·02/0·05	0·008	0·11	0·09
G	3·25	1·5	0·30	0·36	0·005	0·02	<0·005	0·004	—	0·47
H	3·25	0·84	0·53	0·58	0·007	0·009	0·002	<0·002	<0·002	0·07

Weldability of Spheroidal-graphite-type Ni-Resist Cast Irons

the results being reported as the number of cracks per section or per unit length of fusion. This had shown that cracking was accentuated by increasing the phosphorus and magnesium contents, and that enhanced chromium and manganese contents were beneficial; manganese being about half as effective as chromium. Indeed, cracking could be eliminated by increasing the chromium content to greater than 3%, although that could not provide a unique solution to the problem, because the associated loss of ductility and toughness and the increased tendency to microporosity would not be acceptable in all instances.

In a further series of tests, simulated reclamation welds were made using 4-mm. dia. 55% Ni/Fe/C MMA electrodes, details of which are given in Table 2. The weld-preparations were 100-mm. long blind-ended grooves, 22-mm. dia. and 11-mm. deep milled in cast plates of Ni-Resist Type D-2, 150-mm. long, 50-mm. wide and 25-mm. thick. Welding was undertaken without pre-heat and the interpass temperatures maintained below 100°C to minimise the width of the carbidic HAZ.

Examination of several polished sections from each test at magnifications up to ×500 and analysis of the observations showed that cracking could be eliminated when the phosphorus contents of the castings were less than 0.025% and the silicon and magnesium contents conformed to the restrictive relationship: $Si\ \% + 75\ Mg\ \% \leq 7.5$

Among the welding consumables used, there was no

TABLE 3: *Effects of section size, restraint, prior heat-treatment, preheat and postweld heat-treatment.*

Test no.	Iron type	Thickness mm	Prior heat-treatment	Imposed restraint	Electrode (Table 2)	Preheat °C	Interpass °C
1	X	25	None	None	D	None	<100
2	Y	25	None	None		None	<100
3	X	25	Annealed**	None	A	None	<100
4	X	25	None	Fillet welded to 100mm MS block	B	None	<100
5	Y	25	None			None	<100
6	X	50	None	None	A	None	<100
7	Y	50	None	None		None	<100
8	X	25 cut from 100	None	None	A	None	<100
9	X	50	Stabilisation	None	A	None	<100
10	Y	50	Annealed***	None		None	<100
11	X	50 cut from 100	None	None	A	350	350
12	X	50	None	None	D	None	<100
13	X	50	None	None	D	None	<100
14	Y	50	None	None		None	<100

* 2hr/650°C, AC. ** 2hr/1000°C, FC to 525°C, AC. *** 3hr/870°C, FC to 525°C, AC.
X = P—0·02%, Mg—0·07%, Nb—0·10%, Y = P—0·04%, Mg—0·04%, Nb—0·17%. Other elements similar in both 'X' and 'Y' irons.

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TABLE 2: (*continued*).

Consumable identity	Cu	Ti	Al	Fe	Ni
A	<0·05	0·9	0·22	44·8	bal.
B	0·10	0·20	0·43	45·9	bal.
D	<0·05	<0·05	<0·05	45·5	bal.
E	<0·05	<0·05	<0·05	45·9	bal.
F	0·09	0·69	0·1	43·1	bal.
G	—	—	—	bal.	23·5
H	0·05	<0·05	<0·05	44·6	bal.

discernable effect on the extent of HAZ cracking if the weld metals and castings were sound.

It was shown also that greater tolerance for silicon, magnesium and phosphorus was possible if the iron contained a small proportion of niobium. Detailed analysis of data obtained from 84 similar tests enabled practical compositional ranges for these elements to be developed for Ni-Resist Type D-2 exhibiting superior weldability. The other elements, nominally constant, were within the following ranges:

Carbon 2.25-2.95%; chromium 1.7-2.5%; manganese 0.9-1.8%; nickel 18-23%; sulphur less than 0.002-0.012%, and nitrogen 0.002-0.007%

Fig. 1 shows the effect of silicon and magnesium on HAZ cracking in low-phosphorus, niobium-free Ni-Resist Type D-2 castings, the relationship being only slightly modified by comparison with that found previously. Fig. 2, also pertaining to castings low in phosphorus, whilst confirming that small additions of niobium extend the tolerance to silicon and magnesium, also shows tolerance to niobium itself is diminished with increase in silicon and/or magnesium contents. The beneficial effect of niobium on tolerance to phosphorus is illustrated in Fig. 3.

From these tests it was possible to select a composition range for Ni-Resist Type D-2, identified as Type D-2W, which is less prone to weld cracking and whilst being more closely controlled than normal specification requirements, can be obtained by most foundries viz:

Ni-Resist Type D-2W: 3.0 max. T.C.; 1.7-2.2 Si; 1.0-1.5 Mn; 2.0-2.4 Cr; 18-22 Ni; 0.05 max. Mg; 0.040* max. P; 0.12-0.17 Nb; Bal. Fe.

* Prefered phosphorus 0.035% max.

However, specification by composition alone may be unduly restrictive in that advantage cannot be taken of the full scope afforded by the relationships deduced from figs. 1, 2 and 3. To prevent castings which are weldable being unnecessarily rejected, the composition should be checked to determine whether it meets the requirements of the following relationships:

a $Nb \leq 0.353-0.032\ (Si + 64Mg)$

b $Nb \geq 0.0286\ (Si + 64Mg - 5.6)^* + 8\ (P-0.025)‡$

‡ Ignore this term if ≤ 0.

*‡ Addition of niobium is not necessary if both of these terms ≥ 0.

The beneficial effect of chromium on weldability demonstrated in the earlier X-weld tests has been confirmed and extended by results obtained from additional reclamation welding trials as is shown in fig. 4. The data obtained were insufficient for inclusion in the above analysis and were treated separately to give the following tentative relationship:

c $Si + 64Mg + 280P - 35Nb - 3.8Cr \not> 4.3$

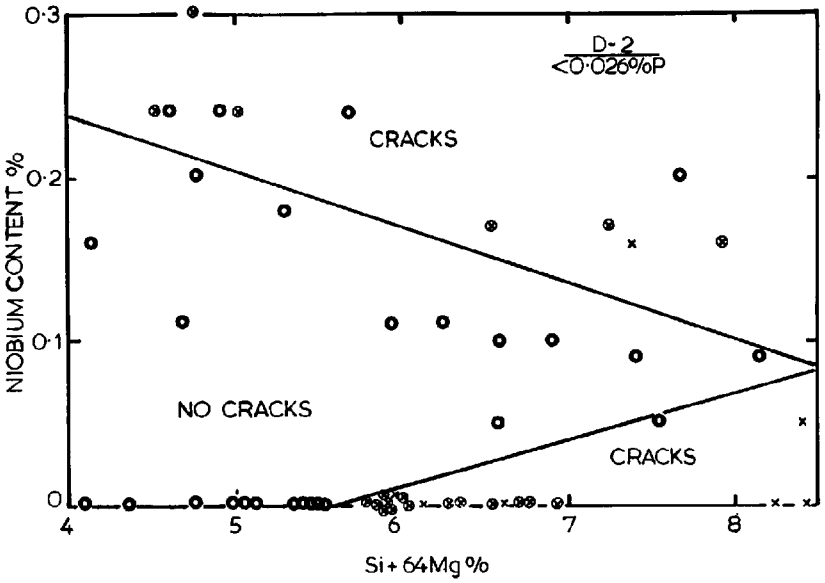
In the above relationships, phosphorus contents below 0.025% are assumed to be 0.025%, because fig. 3 indicates that decrease below that level has no further beneficial effect on weldability. Also, the minimum chromium content which will permit weldability at low levels of silicon and phosphorus and at the highest permissible level of niobium has been tentatively fixed at 1% (fig. 4) because above this level the volume fraction of chromium carbide increases rapidly in the parent casting[5] and a minimum volume fraction of the carbide appears to be

TABLE 3: (*continued*).

Test no.	Buttering	Postweld heat-treatment	HAZ soundness
1	None	Stress-relieved*	Few cracks (<0·5mm)
2	None	Stress-relieved	No cracks
3	None	None	No cracks
4	None	Stress-relieved	Cracks
5	None	Stress-relieved	Slight cracks
6	None	None	Cracks
7	None	None	Slight cracks
8	None	None	No cracks
9	None	Stress-relieved	Cracks
10	None	Stress-relieved	Slight cracks
11	None	Stress-relieved	Cracks
12	Buttered	None	No cracks
13	Buttered	Stress-relieved	Few cracks (<0·5mm)
14	Buttered	Stress-relieved	No cracks

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Weldability of Spheroidal-graphite-type Ni-Resist Cast Irons



2 Effect of niobium on HAZ cracking in groove welds made in low-phosphorus iron D-2.

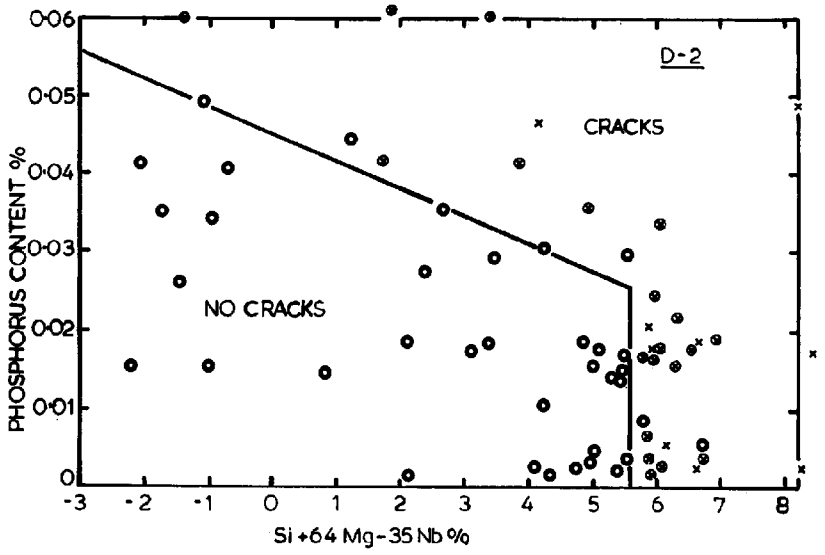
necessary for weldability. Further, it is assumed that the relationships **a** and **b** above are valid in conjunction with **c**.

Included in the tests leading to the construction of fig. 4 were niobium-containing castings of the s.-g. Ni-Resist D-2B type based on the then estimated limits of a practical compositional range for weldability, and the range finally derived is as follows:

Ni-Resist Type D-2BW: 3.0 max. T.C.; 1.5-2.5 Si; 1.0-1.5 Mn; 2.7-3.3 Cr; 18.0-22.0 Ni; 0.05 max. Mg; 0.040 max. P; 0.12-0.17 Nb; Bal. Fe.

As with Ni-Resist Type D-2W, specification by means of compositional range alone may be unduly restrictive in that economic and technical advantages would not be taken of the full scope afforded by the weldability relationship; also weldable castings could be rejected unnecessarily.

Whilst no similar relationships controlling the weldability of the chromium-free grades of Ni-Resist, viz. types D-2C and D-2M, have been deduced in support of the earlier work,[7] reasonable success has been obtained with a version of Ni-Resist Type D-2W low in chromium and phosphorus (composition E in fig. 4). Although outside the specification for Ni-Resist Type D-2C, the iron exhibited satisfactory mechanical properties, viz:

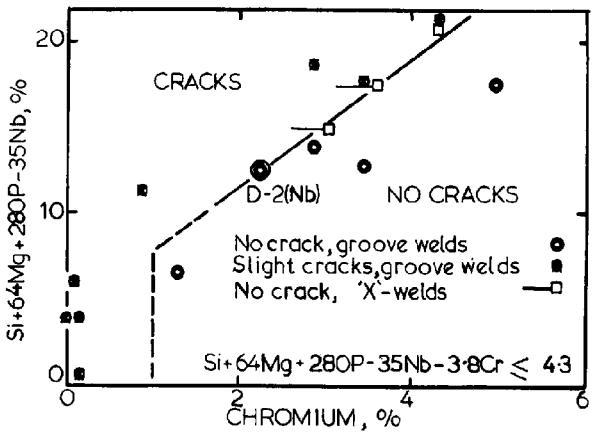


3 Effect of phosphorus on HAZ cracking in groove welds made in niobium-containing iron D-2.

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Weldability of Spheroidal-graphite-type Ni-Resist Cast Irons

	Cr %	Tensile properties 0·2% proof stress (N/mm²)
Ni-Resist Type D-2C Composition E	0·5 max. 1·31	210 231–255
Tensile properties *cont.*		
UTS (N/mm²)	Elong. %	Impact toughness CVN (Joules)
440 437–467	25 22–23	24 27–29

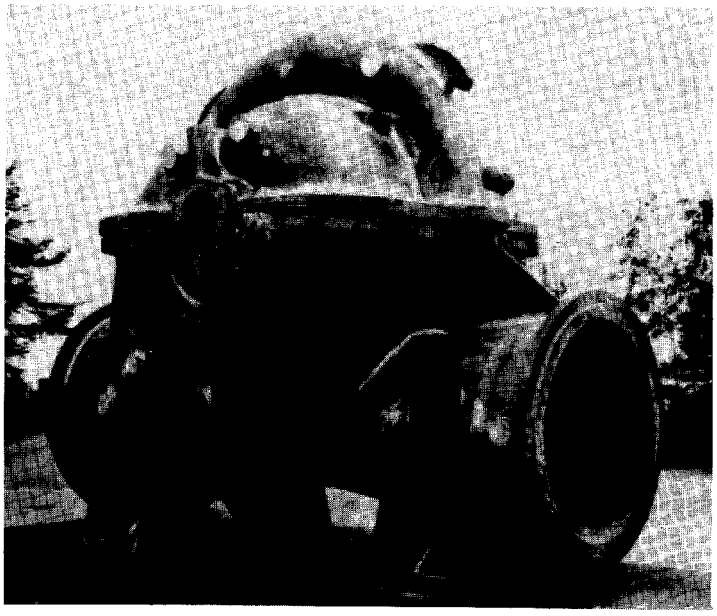


4 Effect of chromium on weldability. Note: if P<0.025, write P=0.025, if Cr<1, write Cr=0).

However, it is doubtful whether a practical compositional range could be derived that would be acceptable to the foundries.

Earlier Inco research had shown that susceptibility of Ni-Resist cast irons to cracking in the weld HAZ is independent of the nickel content of the casting. It is therefore possible from fig 4 to suggest compositional ranges for improving the weldability of the higher nickel-containing grades of s.-g. Ni-Resist iron viz. Types D-3, D-4, D-4A and D-5B assuming that the enhanced nickel contents, and the high silicon content in some instances, do not influence the relationship governing the maximum amount of niobium to be added. Readers are warned, however, that no welding tests have been undertaken with these higher nickel grades nor have any tests been undertaken to determine whether the mechanical properties both at ambient and elevated temperatures will still comply with specification requirements. Correspondingly, if the above assumption is valid, it is unlikely that HAZ cracking would be eliminated from the low-chromium grades, Type D-3A and D-5, by the addition of niobium.

Commercial production of the niobium-containing Ni-Resist Type D-2W is established in both the UK and Germany and in the latter country a tentative specification covering this grade has been issued: GGG-NiCrNb 20 2-Werkstoff Nummer 07659. The weldability of Type D-2W castings produced commercially has been shown to be satisfactory in reclamation trials, and a sound positional

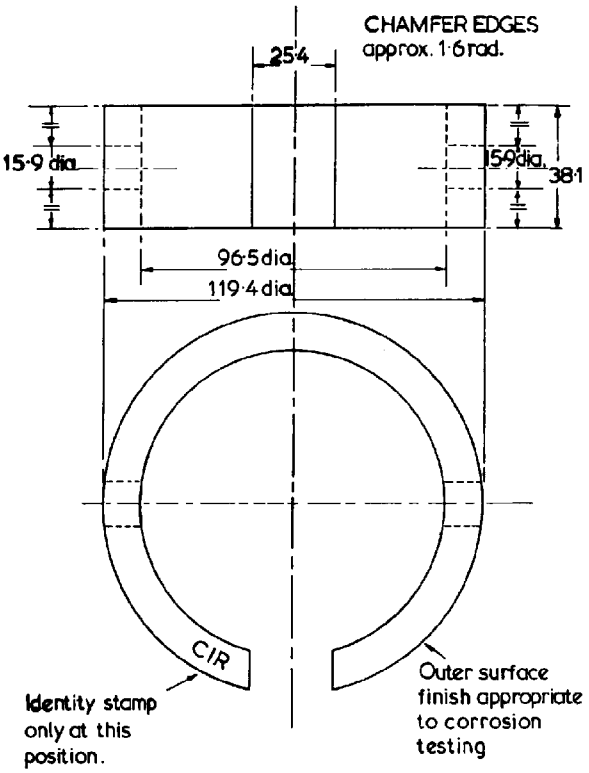


5 Pump Casing in Ni-Resist Type D-2W —total weight, 6,400kg.

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butt-weld has been made with the MMA electrode 'D' (Table 2) to demonstrate potential for assembly by welding.

Fig. 5 illustrates a typical pump casing being made by a German foundry in Ni-Resist Type D-2W for ultimate installation in a Saudi Arabian desalination plant. The castings comprises two castings, the upper weighing 2,100kg and the lower 4,300kg.



6 'C'-ring stress corrosion specimen.

Effects of Section Size, Restraint, Prior Heat-treatment, Preheat and Post-weld Heat-treatment

For some applications, castings are stress-relieved to provide adequate dimensional stability for machining or diminish any tendency to stress corrosion cracking in service; also, they may be annealed to improve ductility and toughness. In those instances, assurance against cracking during post-weld heat-treatment would be sought. The extent of HAZ cracking during welding and subsequent heat-treatment would depend on the weld stress level as governed by factors such as casting thickness, restraint, preheat and heat input. Metallurgical features having potential relevance to cracking, for example, segregation and cell size, would be expected to change with section thickness. Further, the repair welding of heat-treated castings to extend service life may be necessary. The effects of a number of those variables were investigated in trials during which blind-ended grooves, 100mm. long by 23mm. dia. by 11mm. deep in cast blocks, 150 by 100 by 25mm. and 150 by 100 by 50mm. and of compositions 'X', and 'Y', were filled by welding. Composition 'X' contained P - 0.02, Mg - 0.07, Nb - 0.10%, and composition 'Y' contained P - 0.04, Mg - 0.04, Nb - 0.17%; in other respects they were similar. Both were weldable according to the criteria previously derived, but only 'Y' conformed to the D-2W composition.

Short Cracks

The results of these tests, recorded in Table 3, show that severely-restrained, as-cast 25mm. thick blocks, and unrestrained, as-cast 50mm. thick blocks whether post-weld stress-relieved or not, contained short cracks in the weld HAZ, and the effect was not eliminated by pre-heat or by stabilisation annealing before welding. Castings to composition 'Y' suffered less than those of composition 'X' which had a lower niobium content. The HAZ in the unrestrained, as-cast 25mm. block of composition 'X' iron was found to contain a few short cracks after the weldment had been stress-relieved, but that of the corresponding 'Y' iron of higher niobium content was crack-free. The absence of cracks adjacent to the weld in an unrestrained, as-welded, 25mm. thick plate which had been cut from 100mm. thick as-cast block of composition 'X' suggests that stress accumulation is more important in crack promotion than the microstructural changes associated with variation in section thickness. "Buttering" of the weld preparation before filling the groove provides a palliative because the weld HAZ would be less heavily stressed during that initial operation. Indeed, when "buttering" was employed only a few small cracks were observed in the HAZ of a stress-relieved weld in a thick block of the casting 'X' low in niobium and none was found in the corresponding 'Y' casting. It would seem that compositions with niobium contents near to the top of the range, i.e. D-2W, are to be preferred when heavily restrained welds and/or post-weld heat treatment are contemplated, in which instances the practice of "buttering" is to be recommended.

The 55% nickel/iron/carbon weld metals, designed originally for use with unalloyed cast irons, have a mean coefficient of thermal expansion of approximately 13.5 x 10^{-6} per °C over the range 20°C - 650°C, such value being less than that of Ni-Resist Type D-2 (circa 18 x 10^{-6} per °C). The weld metal proof stresses generally equal or exceed those of the parent castings, so that cooling after so-called stress relieving in principle could re-impose tensile stresses of proof stress magnitude on the weld HAZ. In practice, no problems have been reported following the use of 55% nickel/iron carbon weld metals and the application of stress relieving or annealing thermal treatments.

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Weldability of Spheroidal-graphite-type Ni-Resist Cast Irons

Effects of Welding Processes and Consumables

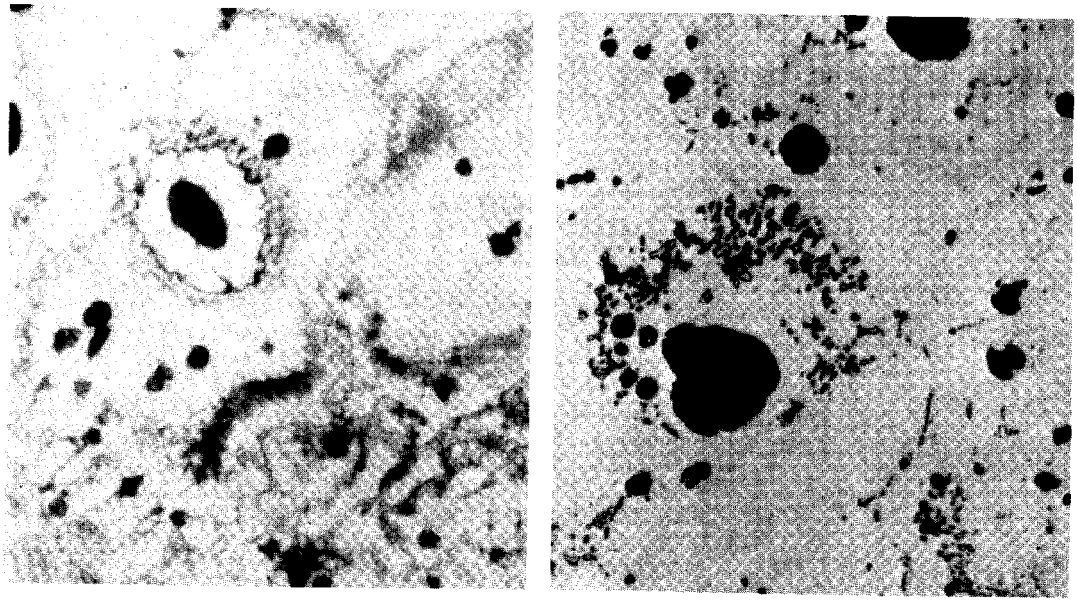
Because of the limitations of the 55% nickel/iron/carbon types of electrodes used for MMA welding of s.-g. Ni-Resist irons, i.e. magnetic characteristics, non-matching coefficients of thermal expansion etc., alternatives were sought, attention being given also to the possibility of semi-automatic welding procedures to facilitate fabrication.

Welds deposited in grooved castings from nickel/copper/carbon (ENiCu-B), bronze and 25-20 stainless-steel MMA electrodes cracked in the weld metal and/or at the fusion boundary. The Ni/C (E-Ni) MMA electrodes were not assessed because manufacturers' data indicated in general that the weld metal strength or ductility would be inadequate. Whereas unalloyed s.-g. iron can be welded successfully by low heat-input MIG techniques[3], using low carbon nickel filler metals, HAZ cracks parallel to the weld fusion boundary are found in s.-g. Ni-Resist castings even in the niobium-bearing composition, Type D-2W. The high solidification temperature of the nickel weld metal is thought to be responsible for the occurrence and configuration of the defect.

A crack-free simulated repair weld was made in a 25mm. thick casting of composition similar to that of Ni-Resist Type D-2W using an Inco-developed experimental MMA electrode 'G' depositing weld metal of s.-g. Ni-Resist composition (Table 2), but, surprisingly, bearing in mind the similarity in coefficients of expansion of the weld metal and the castings, no diminution in HAZ cracking was found in niobium-free cast blocks similar in size and compositions to those which cracked when welded with a 55% nickel/iron/carbon electrode.

The operability, weld metal soundness or detailed mechanical properties associated with certain commercial 55% nickel/iron/carbon MMA electrodes evidently have been of secondary consideration, owing perhaps to development originally for the downhand reclamation and repair-welding of preheated grey cast iron. Such features were considered to be of greater importance in welding the tougher, more ductile, s.-g. Ni-Resist iron for which on-site, positional welding capability without high pre-heat was also sought; another consideration being the need to diminish weld metal defects such as angular flux inclusions which had been observed to induce slight cracks in the adjacent HAZ when the composition of the casting indicated marginal weldability. Positional "bead on casting" tests, used initially to assess the operability of a number of commercial MMA electrodes 3.25mm. dia., were followed by horizontal-vertical, upward-vertical and overhead welds made with selected electrodes in truncated grooves which had been machined in cast plates; the weldments being examined radiographically. Electrode 'D' (Table 2) exhibited outstanding operability and its deposits contained the least number of flux inclusions and pores. Additionally, in making butt-welds, narrower preparations, 60-70° included angle, are tolerated by the electrode 'D'

7 Weld heat-affected zone of niobium-containing 'Ni-Resist' D-2; a (left) as welded, × 500, b (right) annealed, × 400



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TABLE 4: *Tensile properties of all-weld-metals, weldments and unwelded castings.*

Consumable or casting identity	Consumable dia mm	Material	Preheat interpass °C	Postweld heat-treatment	Tensile properties at 20°C (mean of two or more tests)		
					0·2% PS N/mm²	UTS N/mm²	El. on 25mm, %
A	4	Weldment	20+100	As-welded	278	370	2
		,,	20+100	Stress-relieved	256	370	3
B	3·25	Weldment	20+100	As-welded	270	413	2
		,,	20+100	Stress-relieved	256	368	1
		A-W-M†	Not known	As-welded	>296	>396	—
		,, †	,,	Annealed	>310	>448	—
D	3·25	Weldment	20+100	As-welded	294	424	>3
		,,	20+100	Stress-relieved	258	425	>5
		,,	20+100	Annealed	227	472	>13
		A-W-M†	Not known	As-welded	>330	>480	—
		A-W-M	20+100	Stress-relieved	280	414	10
E	3·25	A-W-M†	150+150	As-welded	331	507	—
F	4·0	A-W-M†	20+c.20	As-welded	430	577	—
G	3·25	Weldment	20+100	As-welded	271	434	4
		,,	20+100	Stress-relieved	285	309	2
		,,	20+100	Annealed	221	449	13
H*	3·25	A-W-M	'low'	None	—	490/590	—
Casting (low Nb) Ni-Resist D-2W	—	Casting	—	None	285	395	—
		,,	—	Stress-relieved	288	366	—
		,,	—	Annealed	250	388	—
Casting type (high Nb) Ni-Resist D-2W	—	Casting	—	None	276	423	—
		,,	—	Stress-relieved	271	395	—
		,,	—	Annealed	250	416	—
S-NiCr 20 2 (AUS 202A)		Casting	—	None	215/250	370/460	—

† Consumable manufacturer's data. * Consumable manufacturer's data:
Weldment testpieces were cut transversely from double-vee butt-welds between 13mm thick cast plates.
Plate composition: Type 'X', for welds.

Weldability of Spheroidal-graphite-type Ni-Resist Cast Irons

whereas included angles of up to 90° were required to avoid slag stringers near the weld fusion boundary when using some of other 55% nickel/iron/carbon consumables to produce butt-welds without preheat and at lower interpass temperatures. Of course, operability and soundness are also welder-dependent.

Mechanical Properties

The mechanical properties of Ni-Resist Type D-2 iron are not adversely affected by the controlled additions to niobium; indeed, proof stress may be enhanced (Table 4). In general, the weld metals and weldments develop satisfactory tensile strength, but, in some instances, weld metal impact toughness (Table 5) did not match that of the unwelded castings, again implying that electrode selection is desirable. The toughness of certain weld metals was improved by increasing the preheat and interpass temperatures or by post-weld annealing. Electrode 'B' deposited weld metal relatively high in silicon and exhibiting low toughness even after annealing, but comment on the effects of composition on toughness is difficult because the analytical techniques employed were not capable of distinguishing between the reactive elements in the form of alloyed constituents and the same elements present in oxides and small slag inclusions. Traces of magnesium and calcium, if soluble would be expected to promote beneficial spheroidisation of graphite; certain other elements if similarly present and in small amounts might stabilise the adverse interdendritic filamentary graphite.

The impact toughness of the weld metal from electrode 'D' which had shown excellent operability was only moderate although probably acceptable for many purposes whereas electrode 'E' which is considered to be suitable only for downhand welding operations exhibited toughness values comparable to those specified for castings in Ni-Resist Type D-2.

Fracture Toughness

Fracture toughness tests have been made on samples cut from castings made of Ni-Resist Type D-2W and also from double-vee and "K"-butt welds made downhand between 150 by 100 by 25mm. plates of similar composition, all of which had been stress relieved at 650°C. Use was made of the most promising of the welding consumables evaluated, namely those identified as 'D' and 'E' in Table 2. The results, summarised in Table 6, indicate that welds made with consumable 'E' are of satisfactory quality whilst those made using electrode 'D' are perhaps only marginally satisfactory when examined in conjunction with the lower level of impact properties obtained. Nevertheless, for many applications, welds made with either consumable may be acceptable.

The load-deflection curves, the appearance of the fractures and the K_{max}/K_q ratio obtained during

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TABLE 4: (*continued*).

Consumable or casting identity	Tensile properties at 20°C (mean of two or more tests)			Fracture location in weldment
	El. on 50mm, %	El. on $5{\cdot}65\sqrt{h}$ %	R.A. %	
A	3	—	7	Weldmetal
	5	—	3	,,
B	6	—	5	,,
	3	—	4	,,
	>6	—	—	—
	>8	—	—	—
D	8	—	9	Parent
	10	—	13	,,
	17	—	15	,,
	—	>15	—	—
	8	—	11	—
E	—	13	14	—
F	—	10	—	—
G	8	—	12	Root defect
	3	—	2	Parent
	15	—	13	,,
H*	—	15	—	—
Casting (low Nb)	12	12	12	—
	8	8	9	—
Ni-Resist D-2W	12	12	13	—
Casting type	14	14	14	—
(high Nb)	12	11	11	—
Ni-Resist D-2W	15	>15	>13	—
S-NiCr 20 2 (AUS 202A)	8/20	—	—	—

these tests showed the "COD" or general yielding approach to the utilisation of this data to be appropriate. On this basis it can be calculated assuming an operational stress of 355N per sq.mm. comprised of the proof stress (as residual stress) and a design stress of about 20% of the tensile strength, that the maximum crack half-length (A_{max}), permissible without risking failure is approximately 13mm. The presence of buried, isolated HAZ cracks of, say, 0.5mm. length is thus insignificant. Indeed, isolated angular casting defects of substantially greater magnitude would be tolerated in most castings although if these appeared as

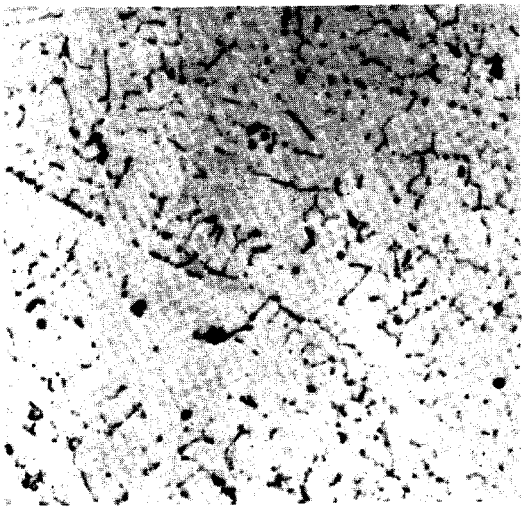


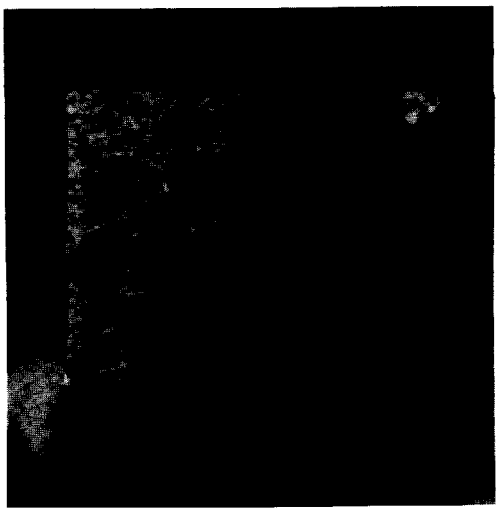




8 Microstructures of castings, × 1,000: **a** (top right) Mg/Si phase—nil Nb; **b** (below) contiguous Mg/Si and Nb/C phases—0.1% Nb, and **c** (bottom right) increase of Nb/C phase—0.3% Nb.

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9a Fracture face, left, in 55Ni/Fe/C weld metal x 5. The dark bands correspond to graphite spheroidisation shown in 9b, right: x 200.

welding-slag stringers at the fusion boundary, they could be significant.

In practice the fracture toughness may be altered unfavourably by section size, corrosive environment, service temperature and mechanical- or thermal-fatigue conditions. The availability of fracture toughness data might allow the use of Ni-Resist Type D-2W where otherwise it would be necessary to select a less-weldable lower chromium grade, such as Ni-Resist Type D-2C.

Stress-corrosion Characteristics

Some tests by Inco had shown that Ni-Resist Type D-2 is susceptible to stress-corrosion cracking in boiling saturated brine under reflux conditions and the cracking becomes apparent within 30 days at stresses approximately 0.2% proof stress of the iron. These findings contributed to a recommendation that castings of Ni-Resist Type D-2 iron should be stress relieved when service in a corrosive environment such as sea water is envisaged. Whilst experience has shown that the cracking rarely occurs in stress-relieved Ni-Resist cast irons, it was considered prudent to ensure that the presence of niobium in Ni-Resist cast iron did not have a detrimental effect. For this purpose a series of C-rings (fig. 6) were produced from annular castings having compositions typical of Ni-Resist Types D-2 and D-2W. All were initially rough machined, a proportion of the D-2W type each having a peripheral semi-circular groove, 6.4mm. radius, which was filled with 55Ni/Fe/C weld deposit from the MMA electrode 'E' (Table 2). Prior to final machining, all the rings were stress-relieved at 650°C. Stressing was effected by bolting across the rings diametrically, the gap closure being calibrated against readings from strain gauges located on the most highly stressed part of the periphery of typical welded and unwelded castings with the rings under known compressive load.

Unwelded rings were stressed to the 0.2% proof stress level and the welded rings to 75% of the 0.2% proof stress of the composite. The lower applied stress for the welded rings was considered advisable because of the residual tensile stresses imposed on the weld heat-affected zone, during cooling after stress relieving, by virtue of the difference between the coefficients of thermal expansion of the casting and the weld metal.

The stressed rings were suspended in boiling saturated common-salt solution under reflux conditions during which time they were examined periodically for cracks, but after 30 days none had been found in any of the castings, weld deposits or weld heat-affected zones. Further tests are in progress in which stressed samples are immersed in an aerated saturated salt solution at 80°C. So far, unwelded specimens of D-2 and D-2W have each withstood stresses of one third and one half their 0.2% proof stresses over half a year, and there is nothing to suggest that the niobium-containing iron is more susceptible to stress-corrosion cracking than the niobium-free material.

A welded specimen of the D-2W iron continues uncracked after five months at a quarter of the proof stress. Normal design stresses are usually one quarter of the proof stress or one fifth of the UTS. In no instances were cracks initiated in the heat-affected zones of welded specimens unless these were associated with areas of gross micro-porosity.

Metallurgical Observations

Previous investigators have indicated a number of factors which may have a bearing on the weldability of s.-g. Ni-Resist cast irons viz:

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Weldability of Spheroidal-graphite-type Ni-Resist Cast Irons

a Lack of hot-ductility in the heat-affected zone, owing to liquation at the cell boundaries, which causes rupture under contraction stresses during cooling[2]. Phases rich in phosphorus and/or magnesium and which would be expected to have melting points below 1,000°C, i.e. significantly lower than that of the solidus of Ni-Resist cast irons have been detected in the cell boundaries.

b Magnesium and silicon have an adverse effect on hot-ductility whereas chromium is beneficial[2]. The amount of chromium-rich carbide in s.-g. Ni-Resist cast irons containing 20% nickel increases rapidly as the chromium content is raised above 1% whilst there is little change in the chromium content of the austenitic matrix when the overall chromium content exceeds 2%. It is known that enhanced cell and interphase surfaces are beneficial in dispersing elements subversive to hot-ductility. Also, there is the possibility that the solubility of subversive elements will be greater in the second phase.

In the present work, variations in silicon, magnesium and niobium levels within the ranges investigated did not reveal any change in either the morphology or the proportion of chromium-rich carbides present nor of the cell size in 25mm thick, sand-cast Ni-Resist Type D-2. Carbides in the HAZ spheroidised slightly and there was some tendency for further carbides to be formed in the vicinity of the existing graphite spheroids (fig. 7a) which when the weldments were subsequently annealed also became spheroidised (fig. 7b), so improving the toughness.

Niobium-free castings contained an elongated hexagonal phase (fig. 8a); this consisting mainly of magnesium and silicon with a little nitrogen. Carbon and oxygen were not found. The volume fraction of cubic niobium-carbide increased with the level of niobium in the casting, some of those carbides being contiguous with the elongated phase (fig. 8b and 8c), owing probably to a common affinity for nitrogen. The presence of both phases unmodified up to the weldfusion boundary, infers low solubility in the matrix and high thermal stability. At high niobium levels, the amount of the elongated phase was much diminished which in principle increases the availability of magnesium to promote HAZ cracks since micro-probe analysis of niobium carbide present shows it to contain significantly less magnesium than the elongated phase. No other niobium-rich phase was found even at niobium contents in the region of 0.3%.

Precipitate clusters near the cell boundaries in the weld HAZ were more evident in the weldable castings containing niobium, and such regions were found to be enriched in Cr, Mn and C. The presence of these carbide precipitates could affect the strength and ductility of the cell boundary at high temperatures.

It has not been possible however to relate fully the weldability of s.-g. Ni-Resist irons to these metallographic observations and further work is necessary to elucidate the beneficial effect of niobium.

The 55% nickel/iron/carbon weld metals from the commercial consumables varied in their tendency to spheroidisation of graphite. In those weld metals displaying predominantly filamentary graphite, thin dark bands at the fracture base corresponded to local spheroidisation on re-heat during successive weld runs (fig. 9). The toughness of the weld metals increased with the extent of graphite spheroidisation, and, given full spheroidisation, improved further

TABLE 5: *Impact toughness values in castings, weld metals and HAZ.*

Consumable or casting identity	Preheat+interpass °C	Location	CVN Impact toughness at 20°C (Joules)		
			As welded	Stress-relieved	Annealed
A	20+100 350+350	WM WM	6, 7, 7 12, 14	6, 7, 8 12, 12, 14	9, 12, 12 14, 15
	20+100 350+350 350+350	WM WM HAZ/FB	6, 6, 6 7, 12, 12 11, 12	4, 4, 5 11, 12 10, 12	5, 5, 10 12, 14, 14 15, 15
B	20+100 350+350	WM WM	5, 5, 5, 5 5, 6, 6	5, 5, 5, 5	5, 5, 5 5, 5, 5
D	20+100 20+100 350+350	WM HAZ/FB WM	10, 10, 11 12, 13 —	10, 11 12, 13 10, 10	11, 11 14, 17 11
E	20+100 150+150 350+350	WM A-W-M WM	— 22+ —	14, 14 — 13, 14	16, 17 — 17, 18
F	20+100 20+c.20 350+350	WM A-W-M WM	— 11, 12+ —	7, 8 — 8, 9	12, 12 — 11, 11
G	20+100	WM	2, 3, 3	4, 4, 5	7, 8, 10
H	100°C	WM	9, 11, 12	11, 12, 9, 10	—
Ni-Resist D-2W (low Nb) NiResist D-2W (high Nb) Ni-Resists D-2		Casting Casting Casting	11, 13, 13 13, 13, 14 12, 12, 12, 14	11, 11, 12 13, 15, 15 12, 14, 14, 15	14, 14, 14 15, 16, 16

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WELDABILITY OF SPHEROIDAL-GRAPHITE-TYPE NI-RESIST CAST IRONS

TABLE 6: *Fracture Toughness of "stress-relieved" butt welds in niobium-modified 'Ni-Resist' D-2.*

Fatigue notch location	MMA electrode	Test-piece thickness mm	KQ N. mm$^{-3/2}$	K max N. mm$^{-3/2}$
Parent casting	Unwelded	50	1,150	1,960
			1,090	1,940
HAZ/fusion boundary (K-butt-weld)	D	23	830	1,660
	E	23	907	1,957
Weld-metal centre-line. (Double-vee butt-weld)	E	23	887	2,185
			845	2,123

* COD at first "event".

Welding details: 'K' welds – 150mm long, 45° bevels. Vertical face buttered prior to welding.
Butt welds – 150mm long, 90° included angle.
All welds – 100°C pre-heat, 100°C interpass temperature.

with spheroid size. A desirable consumable would have the operability and weld metal soundness associated with the electrode 'D' and would deposit weld metal giving at least the toughness of that derived from electrode 'E', without preheat and at low interpass temperatures.

Summary

1 Commercial production of Ni-Resist Type D-2 exhibiting improved weldability has been established, its weldability confirmed and fabrication potential demonstrated.

2 Compositions for Ni-Resist Types D-2 and D-2B irons exhibiting improved weldability have been defined by relationships between the silicon, magnesium, phosphorus and niobium contents and practical ranges are suggested for both grades.

3 For maintenance of weldability with increase in section size, increased restraint, and/or the imposition of post-weld heat treatment, the magnesium content of the modified iron should be restricted and the niobium content enhanced, as in the suggested practical range. Also, "buttering" of the weld preparation is recommended.

4 The mechanical properties of the niobium-modified D-2 irons are satisfactory, and welds made with commercial 55% nickel-iron-carbon consumables develop adequate strength, although consumable selection is required, to ensure acceptable weld metal toughness. The last-mentioned has been related to weld-metal microstructure. Given low heat input

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TABLE 6: (*continued*).

Fatigue notch location	K max KQ	COD max mm	0·2% P.S. N/mm²	Fracture path
Parent casting	1·7	0·26	290	Parent material
	1·8	0·26		Parent material
HAZ/fusion boundary (K-butt-weld)	2·0	0·26		Sheared into parent material
	2·16	0·35		Sheared into parent material
Weld-metal centre-line. (Double-vee butt-weld).	2·46	0·29* 0·40	ca. 330	Weld metal. Crack 'led' by root run
	2·51	0·47		Weld metal initially, then deflected partly into parent material by weld metal defect in root run.

and moderate weld metal toughness, the HAZ-fusion boundary toughness is also satisfactory.

5 Electrodes suitable for downhand repair, reclamation and fabrication in moderate section thicknesses have been found. Further development is necessary, however, to provide a welding consumable which shows an entirely satisfactory combination of operability, weld metal toughness and soundness.

6 Although attempts have been made to relate the weldability of Ni-Resist Type D-2 iron to the effects of composition on micro-structural changes, further elucidation of the beneficial effect of niobium is required.

Acknowledgments

The authors are indebted to Inco Europe, Limited, for permission to publish this paper and to present and former colleagues who have worked on this subject. The paper is based on, and updates, one previously presented[4] in the UK to a conference organised by the Welding Institute in 1976.

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AWS WCI 85 ■ 0784265 0007114 6 ■

The metal arc welding of aluminium alloyed-calcium inoculated grey cast iron at preheat temperatures lower than 200° C.

by Henk de VRIES, Ing., F. Weld. I. (*)

I. — INTRODUCTION

The metal arc welding of non-preheated grey cast irons has been a controversial subject as the formation of a ledeburitic zone at the fusion line could not be entirely avoided.

The development of special electrodes and welding techniques were mainly based on:

a) increasing the ionization in the welding arc by addition of strontium, lithium, caesium or barium to reduce the total heat input for welding
b) the use of weld deposits having a very low hot yield strength to absorb all the welding stresses
c) peening of weld beads to remove tension stresses.

The invention of an aluminium alloyed grey cast iron [3] characterized by high mechanical strength and ductility and slight tendency towards ledeburitic solidification was the starting point of a new investigation into the weldability of grey cast iron.

As these new Al-cast irons are distinguished from the conventional grades, in which silicon acts as a graphitizer by an almost complete replacement of silicon by aluminium [4-5], also filler materials based on that principle were used in our welding trials.

For comparison, commercial available filler materials and a conventional silicon alloyed cast iron base metal were included in the experiments.

II. — BRIEF DESCRIPTION OF THE ALUMINIUM ALLOYED-CALCIUM INOCULATED CAST IRON (3-4-5)

In order to produce this type of grey irons, silicon — the graphitizing element of conventional gray irons — is replaced by the graphitizing element aluminium in combination with calcium as inoculant.

The chemical composition varies between:

C	2-4 %	Si	0-1 %	S	0-0.05 %
Al	1-3 %	Mn	0-0.7 %	P	0-0.1 %

Aluminium strongly retards ferrite formation upon eutectoid transformation. After inoculation, small amounts of ferrite may be found in contact with the graphite flakes. These graphite flakes are relatively short and their distribution is homogeneous. This is mainly caused by the number of eutectic cells which is as high as or even more than 2500/cm².

As a result, the useful range of tensile strengths is from 30-55 kg/mm² with an impact modulus from 9-16 kg . cm/cm³. The relation between Brinell hardness and tensile strength is given by HB = 100 + 3.1 TS.

In bending tests on bars of ϕ 20 × 400 mm, the deflection at rupture is in average 15.7 mm.

By increasing Mn and Cu, the cast iron can be made self hardening in the as cast state.

The extremely low tendency to ledeburite solidification makes this cast iron very suitable for continuous casting and gravity die casting in metallic moulds.

III. — WELDABILITY OF ALUMINIUM ALLOYED CAST IRON IN COMPARISON TO CONVENTIONAL SILICON ALLOYED CAST IRON

As the remarkable good properties of aluminium alloyed grey cast iron come only into existence when the iron is inoculated with calcium, remelting only will deteriorate the material. It is necessary to inoculate the material by each melting operation.

(*) Research Manager of the Belgian Institute of Welding.

Source: Revue de la Soudure, 29(3), 1973 (Belgian Institute of Welding)

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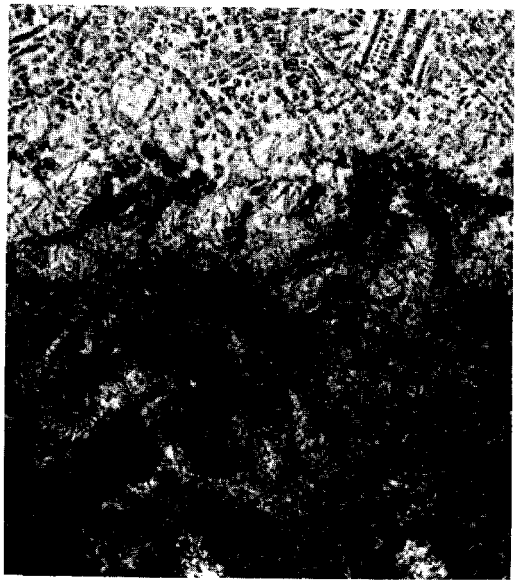
Fusion welding processes without the use of filler material or fluxes containing the inoculant will result in ledeburtic welds, see Fig. 1. For this reason autogenous oxy-acetylene or TIG welding is very unpractical.

However a number of metal arc welding processes are very suitable as the inoculant can easily be put into the coating of an electrode, the core of a wire or in the welding flux. Moreover, in most electrode coatings enough calcium is already present as $CaCO_3$ or CaF_2 to ensure inoculation.

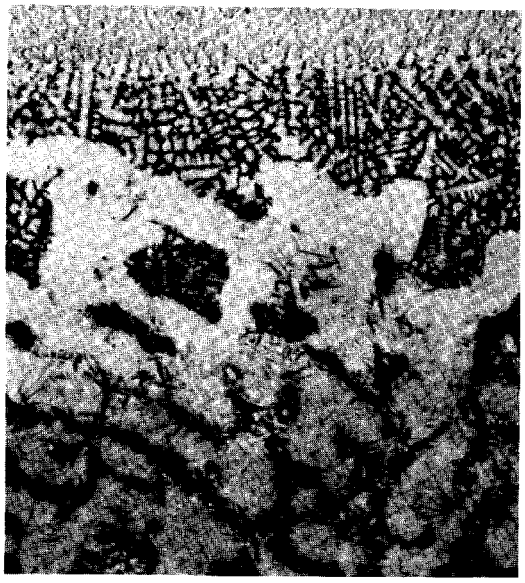
In the heat affected zone all good properties still exist as no melting occurs.



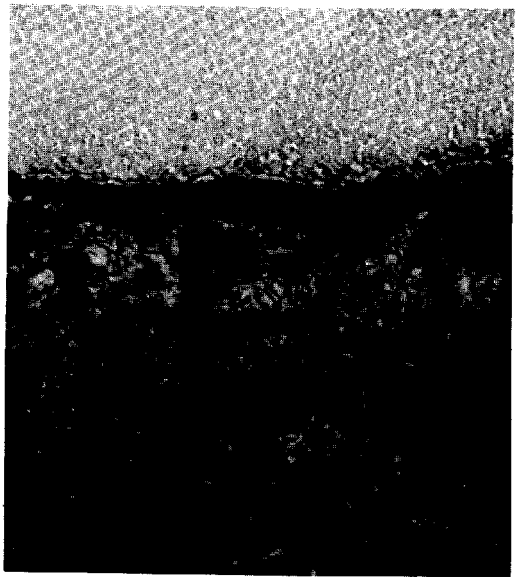
a) Al-cast iron



b) Si-cast iron



c) Al-cast iron + inoculant powder on the surface



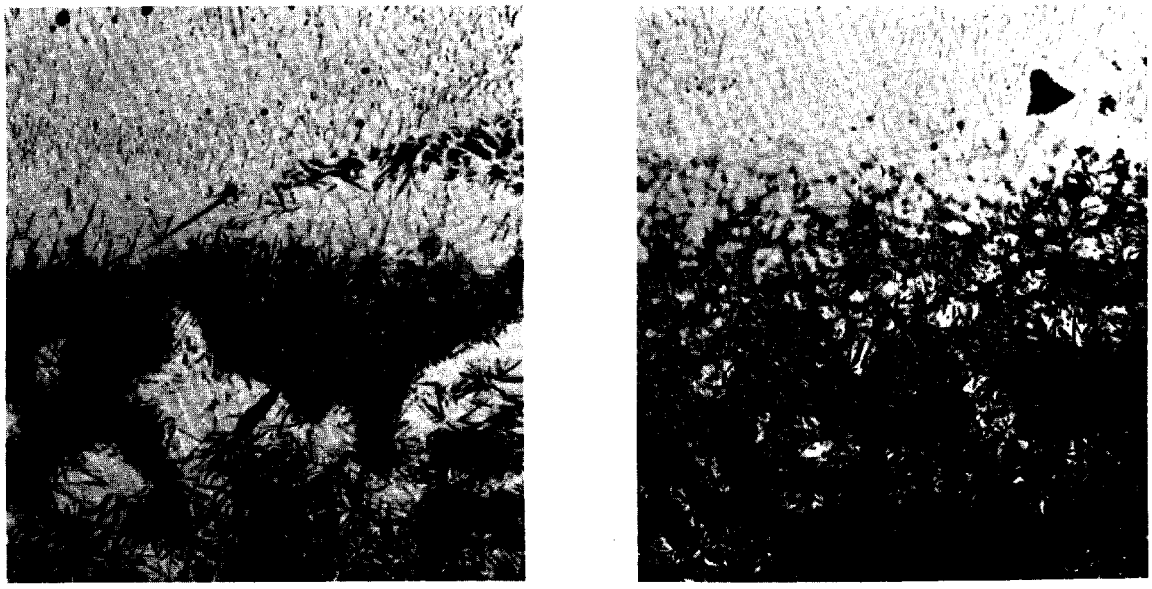
d) Si-cast iron + inoculant powder on the surface

Fig. 1. — Autogeneous TIG welding trials (Nital etch, 135 ×)

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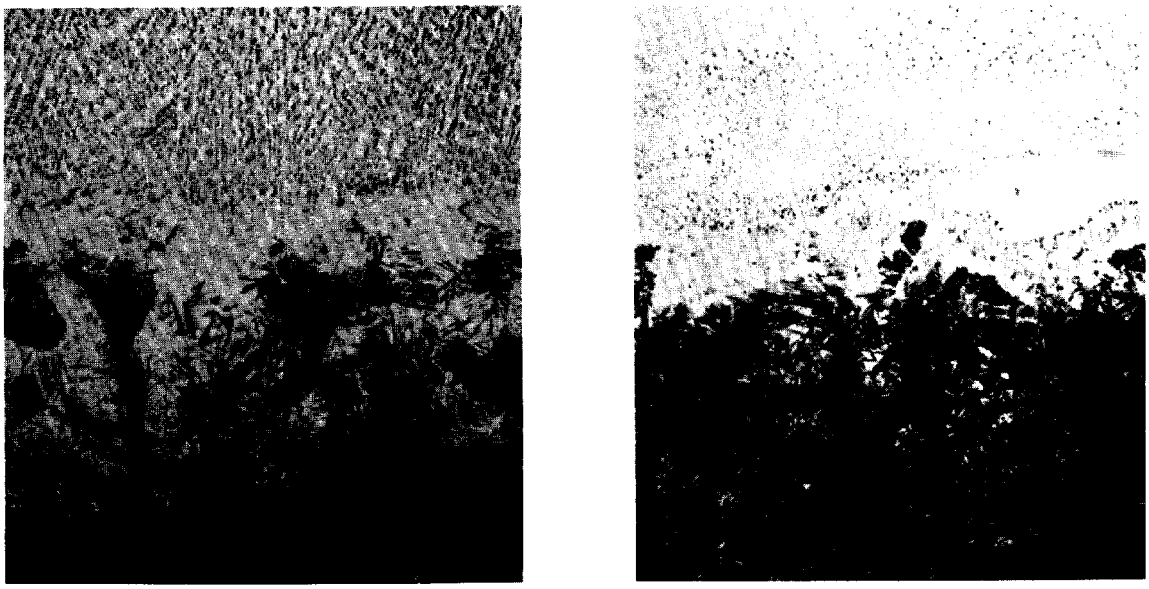
In welding trials with commercial available filler materials no ledeburite could be formed at the fusion line and in the heat affected zone by stick electrode manual electric arc, CO_2 and MIG welding, using the Al-cast iron as base material. Using the same filler materials, welding processes and welding parameters but Si-cast iron as base material resulted in a ledeburitic zone at the fusion line, see Fig. 2, 3, 4, 5 and 6.

Special attention must be drawn to the results obtained with CO_2 welding with a low carbon steel wire.



a) Al-cast iron b) Si-cast iron

Fig. 2. — Manual electric arc welding with a basic 19 % Cr 10 % Ni electrode (Nital etch, 135 ×)



a) Al-cast iron b) Si-cast iron)

Fig. 3. — Manuel electric arc welding with a basic Nickel electrode (Nital etch, 135 ×)

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Unless the carbon content of the weld metals were increased considerably by dilution with the base metals and no extra inoculant was used, still the weld deposits contained no ledeburite. The Si-cast iron showed however a considerable amount of ledeburite at the fusion line. The Al-cast iron none at all.

Welding trials to prove the effect of inoculation were also carried out with cast rods of 5 mm diameter made of 2.76 % Al-alloyed cast iron with 3.8 % carbon, manual electric arc welded through a layer of commercial welding fluxes with and without an inoculant.



a) Al-cast Iron



b) Si-cast Iron

Fig. 4. — MIG welding with a 60/40 NiFe filler wire of 0,8 mm ∅ (Nital etch, 135 ×)



a) Al-cast iron



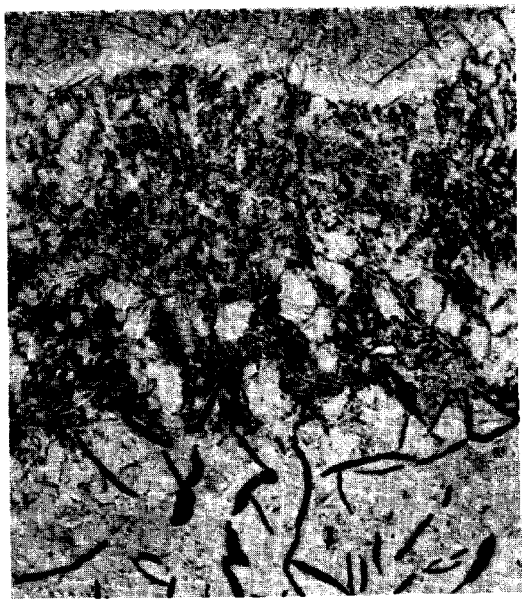
b) Si-cast iron

Fig. 5. — MIG welding with a 61/23/14 NiCrFe filler wire of 1.0 mm ∅ (Nital etch, 135 ×)

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a) Al-cast iron welded at 20° C



b) Si-cast iron welded at 20° C



c) Al-cast iron welded at 200° C



d) Si-cast iron welded at 200° C

Fig. 6. — CO_2 welding with low carbon steel wire of 0.8 mm Ø (Nital etched, 135 ×)

The welds made *without* the addition of an inoculant to the fluxes showed ledeburitic weld deposits.

The welds made *with* the addition of an inoculant showed practically never any ledeburite.

Due to the relatively high electrical resistivity of these cast welding rods, only very short arcing times could be used to prevent overheating of the electrode.

For this reason further work was done with a number of experimental cored wires and normal commercial available welding fluxes.

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IV. — SUBMERGED ARC WELDING OF GREY CAST IRONS WITH CORED WIRES

For reasons of economy the inoculant was only put into the cored wires and not in the fluxes. In principle the results are in general the same when the inoculant is put into the flux.

TABLE I
Composition of fluxes

	A	B	C	D
SiO_2	6	3	45	–
Al_2O_3	20	18	2	30
TiO_2	3	–	–	10
MnO	2	–	–	5
CaO	12	7	13	–
CaO_3	–	–	–	10
CaF_2	55	57	–	45
MgO	2	–	26	–
Na_3AlF_6	–	–	14	–
K_2O	–	2	–	–
Cr	–	13	–	–

The wires 1 to 12 were 3.2 mm ϕ and wire 13 was 2.0 mm ϕ and the grain size of the powders used in between 0.1 and 0.3 mm. For the compositions of the fluxes, see Table I, and for the wires see Table II.

With all the wire and flux combinations weld runs of approximately 100 mm long were deposited using 250-300 Amps at 26-32 Volts and a welding speed of 20 cm/min.

The weld runs were visually inspected and sectioned for microscopical examination. The results are given in Table III in which the following symbols were used:

A = austenite
G = graphite
C = carbide particles
B = bainite
M = martensite (see Fig. 7b)
L = ledeburite (see Fig. 7c)
or carbide network (see Fig. 7a)
U = unacceptable shape
P = pores
S = cracks

in the following order : heat affected zone/weld metal/ weld appearance.

TABLE II
Composition of the cored wires in gr.

	1	2	3	4	5	6	7	8	9	10	11	12	13
Iron	2.55	2.52	2.19	2.01	–	–	2.72	–	–	9.13	6.80	6.08	–
Cast iron	6.12	6.06	6.06	6.03	–	–	3.40	–	–	–	–	–	–
Graphite	–	–	–	–	3.40	2.72	–	6.80	5.10	–	–	–	9.00
Aluminium	1.02	1.01	1.01	1.00	2.72	2.72	1.02	2.72	2.72	3.22	3.22	3.22	–
Al/Mg 70/30	2.89	2.86	2.86	2.85	5.61	5.61	2.89	5.44	5.44	–	–	–	–
Al/Mn 50/50	1.19	1.18	1.18	1.17	–	–	1.19	–	–	–	–	–	–
Al/Ca 50/50	0.34	0.67	1.01	1.34	0.17	0.17	–	0.34	0.34	0.72	–	0.72	–
Al/Fe 50/50	–	–	–	–	–	–	–	–	–	–	–	–	12.00
Fe/Si/Al/Ca 40/25/20/15	–	–	–	–	–	–	–	–	–	–	–	–	3.00
Fe/Mn/C 50/45/5	–		–	–	1.02	1.02	–	1.02	1.02	–	–	–	–
Nickel	–	–	–	–	–	–	2.89	–	–	–	3.04	3.04	–
CaF_2	16.66	16.49	16.49	16.41	18.36	19.04	16.66	14.96	6.66	17.54	17.54	17.54	–
$CaCO_3$	–		–	–	1.36	1.36	–	1.36	1.36	–	–	–	–
MgO	2.30	2.36	2.36	2.34	1.36	1.36	2.38	1.36	1.36	2.51	2.51	2.51	–
Potassium bichromate	0.85	0.84	0.84	0.84	–	–	0.85	–	–	0.89	0.89	0.89	–
Ratio powder/ strip	0.34	0.34	0.34	0.34	0.34	0.34	0.34	0.34	0.34	0.34	0.34	0.34	0.24

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TABLE III
Results of welding trials

Wire n°	Base material	Welding fluxes			
		A	B	C	D
1	Al-cast-iron Si-	M/M/UPS ML/AC	M/ACM/P ML/AM/UPS	M/AM/S ML/AML/US	ML/AM/U ML/ACM/UP
2	Al- Si-	M/AM/P ML/AC/P	ML/AM/U ML/A/UPS	M/AM/S ML/AML/UPS	ML/AM/U ML/B
3	Al- Si-	M/M/UP ML/ACM/UPS	M/AM ML/AM/PS	M/AM/S ML/AM/UPS	ML/AM/UP ML/AM/U
4	Al- Si-	M/AM/P ML/M/P	M/AM/UPS ML/AC/UPS	M/AM/S ML/ACM/PS	ML/AM/U ML/AM/U
5	Al- Si-	M/AM/P ML/M/P	M/ACM ML/AC/P	M/AL/P ML/ACM/PS	M/AC/P ML/AML/UPS
6	Al- Si-	ML/AM/UP ML/M/PS	M/ACM/S ML/AL	M/AM/U ML/ACM/PS	M/AGL* ML/AML/P
7	Al- Si-	M/AM/UP ML/ACM/UPS	M/AML M/AC/PS	M/AC/US M/AM/PS	ML/AM/UP ML/B/UP
8	Al- Si-	M/AL ML/ML/U	ML/AL ML/AC	M/AM ML/ML	ML/AC/P ML/L/P
9	Al- Si-	M/AM ML/ML	ML/AL ML/AC	M/AC/UPS ML/ML/P	M/ACM/U ML/AC/UPS
10	Al- Si-	ML/AM/P ML/AM/P	M/ML/U M/AC/UP	M/ACM/S M/ACM/PS	M/GB* M/AM
11	Al- Si-	M/AM/UP ML/M/P	M/ACM/P ML/AC	M/ACM/PS ML/AM/US	M/AM/UP ML/B/P
12	Al- Si-	M/A/P ML/AM/P	M/AM/P M/AM/PS	M/AM/P ML/ACM/UPS	M/ABM/U ML/AM
13	Al- Si-	M/GB/UP* ML/GB/UP*	M/GBL* ML/AL/UP	M/GCB** ML/AGB/P*	ML/AML/U ML/AM/UPS

* promising results
** the best result, used for welding a test plate

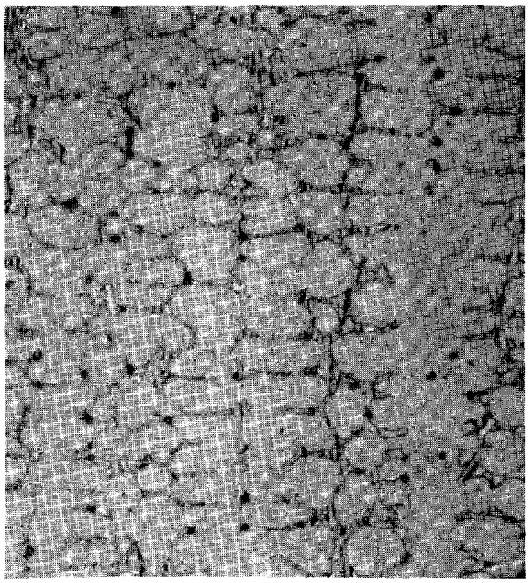
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Typical examples of the microstructures found are given in Fig. 7.

From the results it can be seen that martensite in the weld deposits gives much more cracking than ledeburite.

Furthermore most wire/flux combinations tend to give retained austenite in the weld deposits.

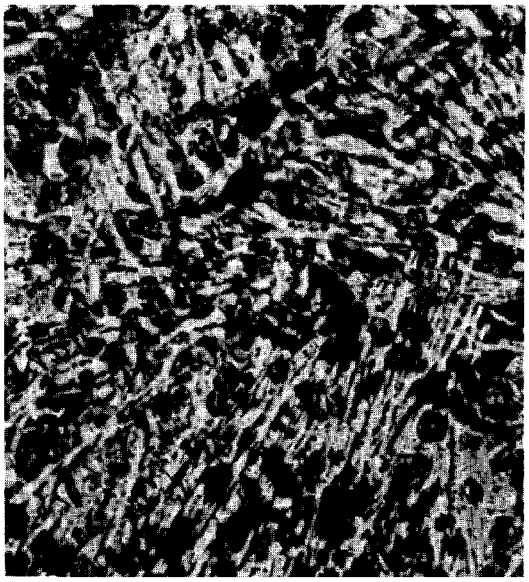
The wires 6-10-13 showed the most promising results. The combination of wire 13 with flux C welded on Al-cast iron showed a grey cast iron weld deposit. This combination was used for welding some test plates.



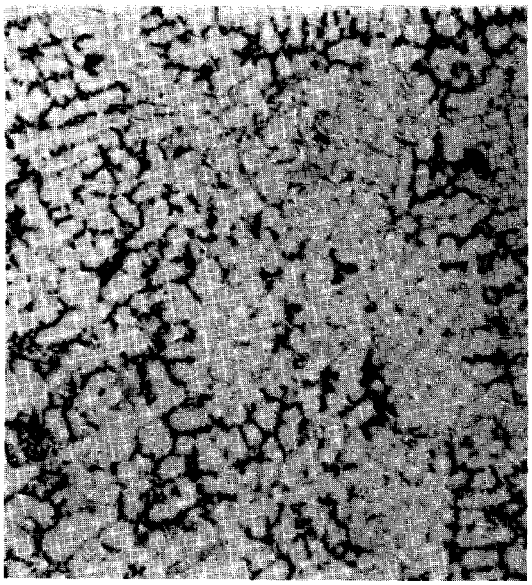
a) Austenite + carbide network (Nital etch, 500 ×)



b) Martensite + carbide particles (Nital etch, 500 ×)



c) Ledeburite (Nital etch, 400 ×)



d) Austenite + graphite + carbide network (Nital etch, 400 ×)

Fig. 7. — Typical examples of microstructures

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V. — WELDING OF TEST PLATES WITH WIRE 13 AND FLUX C

A plate of 300 mm long and 25 mm thickness was welded with a V weld preparation. During welding shrinkage cracking occurred after depositing the last runs. By using a X weld preparation and preheating to max 150°C, this shrinkage cracking could be avoided. Two test plates were welded and inspected by radiography. No significant weld defects were found.

From the welded plates cross sections were taken for microscopic examination and machined cylindrical tensile bars for mechanical testing. A general view of the weld cross section is given in Fig. 8a.

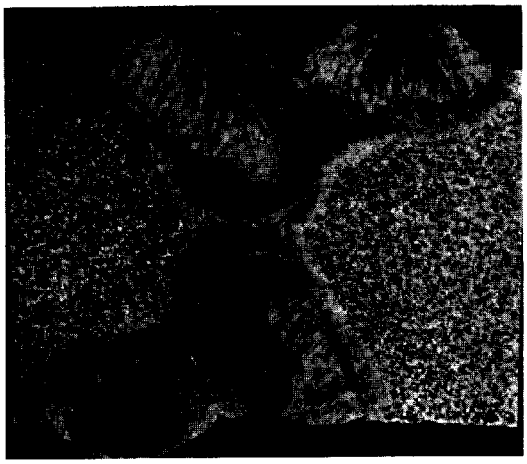


Fig. 8a. — Cross section of submerged arc weld made with wire 13 and flux C and Al-cast iron base material (Nital etch, 2 ×)

The heat affected zone was absolute free of ledeburite, see Fig. 8b, and the structure of the weld deposit comparable to grey cast iron. Depending on the place lamellar or spherical graphite is formed due to reheating of subsequent layers, see Figs. 8c and 8d.

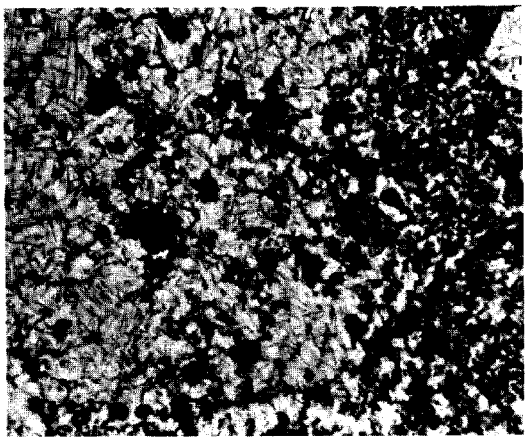


Fig. 8b. — Fusion zone of submerged arc weld made with wire 13 and flux C and Al-cast iron base material (Nital etch, 500 ×).

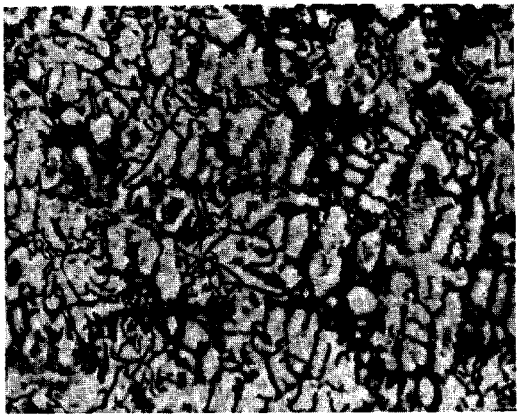


Fig. 8c. — Weld material along fusion line of submerged arc weld made with wire 13 and flux C and Al-cast iron base material (Nital etch, 500 ×)

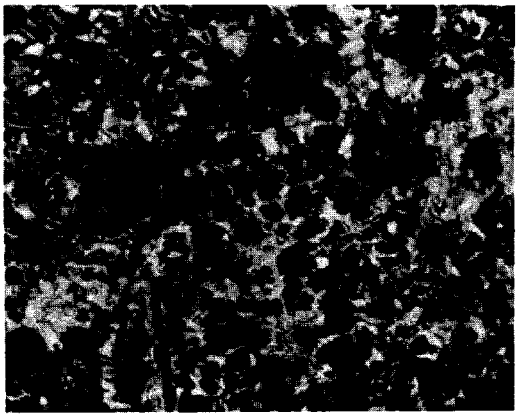


Fig. 8d. — Weld material in center of last runs of submerged arc welding made with wire 13 and flux C and Al-cast iron base material (Nital etch, 500 ×)

The hardness of the base material was 180 Brinell, of the weld 250 Brinell.

The tensile strength of the base metal was 30 kg/mm² and a test bar taken across the weld fractured far outside the weld region. The tensile strength of the weld metal was 38 kg/mm².

VI. — CONCLUSION

Contrary to silicon alloyed grey cast iron, the welding of aluminium alloyed-calcium inoculated grey cast iron by metal arc processes results in ledeburite free

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fusion lines and heat affected zones and with the right combinations of wire and flux or coating in grey cast irons weld deposits with good mechanical properties [6].

Welding of silicon alloyed cast iron with the aluminium and calcium alloyed wire and flux combination can result in grey cast iron weld deposits but might show some ledeburite at the fusion line due to silicon taken up from the base material.

Due to the high amount of aluminium in the experimental aluminium and calcium alloyed wires, open arc welding was possible with all the wires but tending to give a high amount of retained austenite in the weld deposits.

VII. — ACKNOWLEDGEMENTS

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V-2

THE EFFECT OF MICRO AND MACRO STRUCTURES ON PHYSICAL PROPERTIES OF CASTINGS

Considering the importance of gray cast irons as an engineering material, relatively little attention has been given to studies on the macro and micro characteristics of gray iron and their effects on the physical properties of the castings. This has been partly due to difficulties arising in the development of a preparation technique adequate for revealing the true unetched and etched structures and in reproducing these structures for illustrative purposes.

It is the purpose of this article to outline and demonstrate the effectiveness of a simple grinding and polishing technique, developed to obtain the true micro and macro structural characteristics of gray irons in both the unetched and etched condition, and to indicate how the observed structures can be recorded for illustrative purposes. The report also presents a preliminary study on the correlation of casting conditions with the tensile properties of Class 50 gray iron and indicates an association of certain rather clearly defined macrostructural ap-

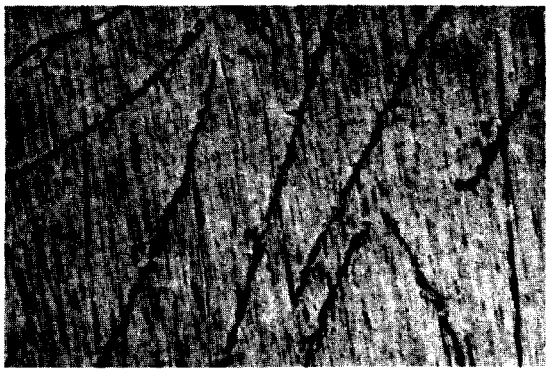


Fig. 1—Gray iron, polished through No. 000 paper and magnified 200 times.



Fig. 2—The same area as shown in Fig. 1 is illustrated here after final polishing, enlarged 200 diameters.



Fig. 3—The same area as shown in Figs. 1 and 2, etched in 4 per cent picrol, enlarged 200 diameters.

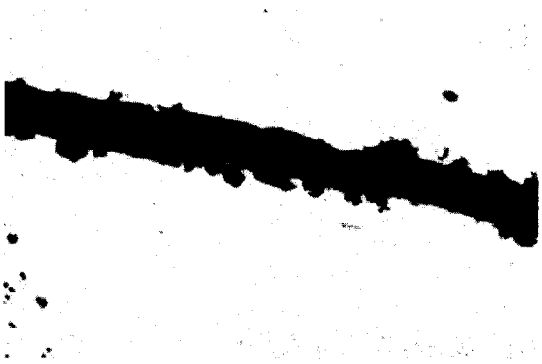


Fig. 4—Graphite flake, unetched, and enlarged 1000 ameters.

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pearances of the cast bars with specific tensile properties.

The sample, cut to convenient size, is ground to a flat surface on a wheel, then ground successively through a series of emery papers: 1G, 1, 0, 00, and then on a 000 paper which has previously been rubbed with cotton to remove any loose grit. Alcohol may be used to moisten this 000 paper, prior to rubbing with cotton, but in either case a dry paper is used for the final grinding operation. At this stage, the scratches on the specimen should be very fine and uniform, as illustrated in Fig. 1, at 200 diameters. The graphite flakes are actually covered with a film of flowed metal which helps to prevent them from being torn out during subsequent polishing procedures.



Fig. 7—This is the same area as shown in Fig. 6, using DI9 for plate development at the same magnification.

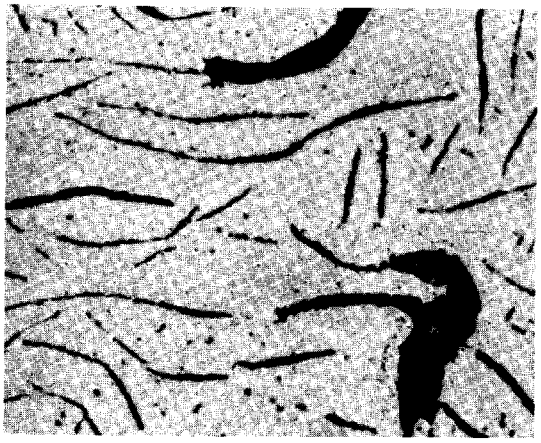


Fig. 5—Shown here are details of graphite flakes, reproduction of which is dependent upon the use of antihalation panchromatic plates, at 100 diameters.



Fig. 8—The etched structure of a high carbon iron brake drum after developing the plate in D76C solution. 600 diameters.



Fig. 6—The etched structure of Meehanite Type GC, obtained after developing in D76C solution, at 1000 magnifications.



Fig. 9—This is the same area as shown in Fig. 8, using DI9 for plate development, 600 diameters.

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The actual polish is obtained in several operations using a "selvyt" cloth, impregnated with magnesium oxide or "shamva" paste. To prepare the cloth for polishing, it is placed on the wheel and saturated with water, after which dry "shamva" powder is rubbed into the cloth until a pasty condition is obtained. No further addition of water or powder is necessary or desirable for completely polishing the sample. The paste accumulating on the edges of the sample, however, should be returned to the wheel. During the polishing operation the specimen should be rotated in a direction opposite to the travel of the wheel.

After final grinding with the No. 000 paper, the specimen is polished on the prepared "selvyt" for a period of ½ to 1 min. with moderate pressure. Many laboratories today prefer to charmand paste and Kerosene. It is then washed thoroughly, dried with alcohol, and lightly etched by a simple dip in a solution of 4 per cent nitric acid in alcohol. The specimen is washed, dried, and repolished for about 30 sec., to remove the light etch. This light etch and polish procedure is repeated from three to five times in order to obtain a scratch-free unetched surface as illustrated in Fig. 2, which shows typical graphite flake distribution in an ordinary gray iron. The complete retention of these graphite flakes is readily noted, and the satisfactory nature of the polish as well as the sharp delineation between graphite and matrix is perhaps more clearly revealed in Fig. 4 at 1000 diameters magnification.

An etched section of the area shown in Figs. 1, 2 and 3 was obtained by use of the conventional picrol solution. This reagent permits better control of depth and uniformity of etch than the 4 per cent nitol solution. This rapid polishing and etching procedure has been utilized in this study on the nature of the graphite flakes observed in ordinary gray irons.

Graphite flakes

When a graphite flake is observed under the microscope, it is light dove gray in color. However, it has usually been found difficult to transfer this tone quality to the photographic plate and print, and it has become customary to note the presence of black graphite flakes in an unetched white matrix or in an etched background of pearlite. Perfect reproduction might be approximated if color photography is used instead of panchromatic type plates, where tone quality alone must serve to give definition.

Various types of wratten filters were used to try to improve the detail and tone quality in the reproduced structure of gray irons. Although the visual appearance of the structure could be slightly improved with certain of these filters, this improvement did not show in the photomicrographs taken. Accordingly, the usual yellow-green filter was used throughout in the preparation of illustrations for this paper.

Further studies with respect to photographic plates, developing solutions, and suitable grade printing papers

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demonstrated that it was practicable to establish conditions which permitted reproduction of structures essentially similar in appearance to those viewed under the microscope.

Consider first the appearance of an unetched microsection at 100 diameters. This magnification is widely used to indicate the size and distribution of the graphite flake, and the photomicrograph invariably consists of black stringers in a white, unetched matrix. All true quality is lost on the photographic plate. By use of a backed plate, however, in which halation is largely eliminated, the actual visual appearance of the properly polished specimen can be transferred to the negative. By use of a suitable paper these qualities can also be transferred to the photographic print. Fig. 5 demonstrates this feature rather clearly and it should be evident that the details observed in the graphite flakes are not dependent on under-exposure or under-development of the print. It may be noted that the matrix or background of the structure is gray rather than white, but it is suggested that this feature in no way detracts from the instructive usefulness of the illustration.

In reproducing etched structures of gray irons at higher magnifications, it is a simple matter to demonstrate the importance of plate developing solution. Figs. 6 and 7 were taken from identical areas of an etched structure at 1000 magnifications using anti-halation Wratten M plates. The former was developed in D76C solution and the latter in solution D19. It must be assumed that optimum printing conditions were employed in each case. In Fig. 6 may be noted the clear details of both the pearlitic structure and the graphite flakes. A similar contrast may be drawn between Figs. 8 and 9 at 600 diameters.

These photomicrographs illustrate the same general features but emphasize more clearly the definition and detail of all constituents, obtained by using D76C solution. The tone quality of both the graphite flakes and the non-metallic inclusions is well brought out along with the details in the pearlite and the scattered free cementite. Fig. 9 is typical of many photographic reproductions of etched gray irons which are familiar, but Fig. 8 presents a much better reproduction of the iron as observed under the microscope. The appearance of the non-metallics in Fig. 8 as contrasted with those in Fig. 9 can also be noted.

The response of this technique to photographic efforts to reproduce micrographs which permit study of the nature of the graphite flake concomitantly with all other structural details is discussed more fully in the next section of the article.

Now that the method of preparation of gray iron has been discussed, attention is directed to a study of the true structural appearance of graphite, especially the flake type, as observed under the microscope.

Frequent references have been made to primary and secondary graphite, but terms are so ill-defined that it is not clear as to what exact significance various authors wish to ascribe to these terms.

Here, it is believed that it has been possible to distin-

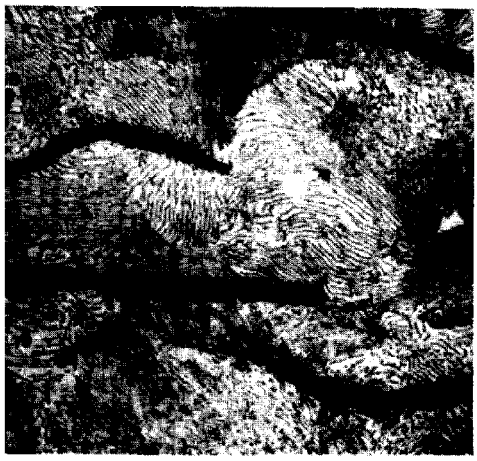


Fig. 10—Etched structure of gray iron showing the simple primary graphite flakes in a pearlite matrix. 600 diameters.



Fig. 11—Etched structure of gray iron showing the simple primary graphite flakes in a pearlite-ferrite matrix. Also can be observed the fine flake graphite network at the ferrite boundary. Enlargement, 1000 diameters.



Fig. 12—The continuous envelope of secondary graphite around the primary flake may be observed in this etched gray iron sample, enlargement 1000 diameters.

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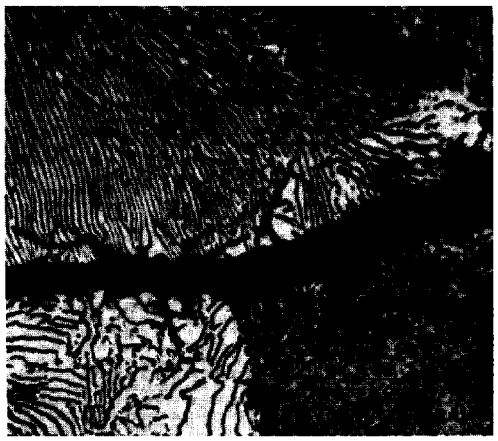


Fig. 13—An etched gray iron sample showing a fine secondary graphite developing from the primary flake. 1000 diameters.

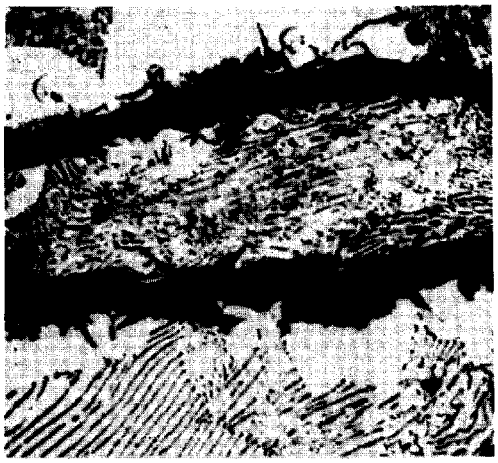


Fig. 14—Another typical combination of primary and secondary graphite in gray iron, enlarged 1000 times.

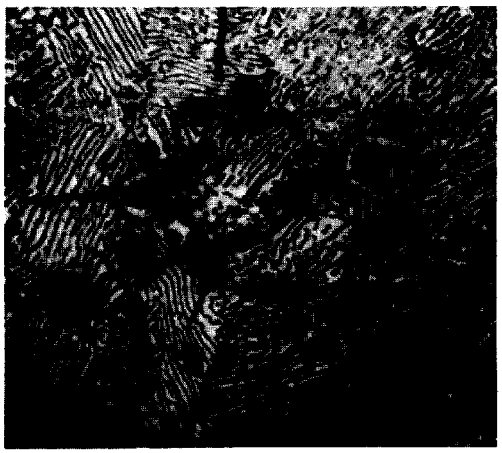


Fig. 15—Etched gray iron, free from primary flake but exhibiting fine secondary graphite development. Enlarged 1000 diameters.

guish between flakes that appear to be homogeneous and which were formed entirely during some definite stage in the initial cooling of the gray iron, and flakes that exhibit an apparent two-stage formation.

Examples of the first class are well shown in microsections taken from Class 50 gray iron castings. Thus, the continuity of structure of the graphite is well illustrated in Fig. 10 from a brake drum casting. Similarly, primary graphite can be found in a gray iron as shown in Fig. 11. This cast iron sample also exhibits a fine graphite network in the pearlite-ferrite boundary.

An example of what may be termed secondary graphite is illustrated in Fig. 12. On either side of the main or initial primary graphite flake has been deposited a continuous regular film or layer of secondary graphite surrounding and following the general contour of the primary flake. While it is not the purpose here to attempt to correlate specifically the stages of graphite formation with the history of cooling of the iron, nevertheless it seems clear that a differentiation must be made between the two stages of graphite flake development.

Another form of this secondary graphite is pictured in Figs. 13, 14, and 15. Figs. 13 and 14 show a combination of both primary and secondary graphite, while Fig. 15 exhibits an area completely free from primary graphite flakes but where secondary graphite can be seen distributed among the pearlite grains.

While the science has not been sufficiently developed to permit a direct correlation between the presence, amount and nature of distribution of what has been termed secondary graphite with physical properties, there appears to be evidence to indicate that melting and casting conditions which prevent the formation of secondary graphite also lead to improvement of physical properties of the gray irons.

Macrostructures of iron

In an effort to ascertain the reason for the profound effect of pouring temperature on the tensile strength of Class 50 iron, studies were made on the changes in macro-structural appearance of the iron as a possible index of properties.

At first, attention was directed to graphite distribution as revealed by low power macroscopy. In these studies it was soon realized that while visual studies could be made satisfactorily using various optical equipment, it was a matter of considerable difficulty to adequately reproduce in photographic form the sort of characteristics that were visually noted. The major problem was found in attempting to reproduce an area of the specimen at low magnification of sufficient size to illustrate significant pattern characteristics and at the same time demonstrate clearly the pattern detail under consideration.

The Bausch and Lomb wide field binocular microscope, using a wide field lamp, affords an excellent means of visual comparison of graphite form and distribution, but this method is not amenable to photographic record. The lamp illuminates a very large field

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for visual study and by means of its ellipsoidal reflecting surfaces throws light, into every recess and eliminates all shadows.

For photographic reproductions of etched sections at magnifications within the approximate range of 1 to 10 diameters, good results have been obtained by use of suitable Bausch and Lomb micro tessars (6½ in., 72 mm., and 48 mm.) with vertical illumination. The illumination is obtained by interposing a plane glass reflector between the lens and the specimens to be photographed. For reproduction of graphite distribution observed in the unetched samples at about 10 diameters, the optimum condition has been found in utilization of the low power equipment provided with the Zeiss Neophot metallograph, using vertical illumination.

Three typical photographs illustrating the appearance of the polished unetched samples of a gray iron taken: (1) From the rim of a truck brake drum in Fig. 16; (2) from an ordinary sample of gray iron in the laboratory in Fig. 17; and (3) from an arbitration bar of Class 50 iron in Fig. 18. These clearly reveal the difference in graphite flake size and distribution. Even at the low magnification of 10 diameters, the "whorl" and "rosette" characteristics of the graphite in Fig. 17 can easily be distinguished from the fine, even, dendritic pattern noted in the sample in Fig. 18.

In white irons the macro characteristics that can be revealed so readily on simple etching are generally familiar. In a study of the effect of casting temperature on the primary microstructure of white cast irons, Norbury presents some beautiful reproductions of the etched macrostructures observed at 15 diameters.

In an attempt to make a similar comparison between pouring temperature and etched macrostructure and to correlate the characteristic effects of pouring temperature on macropattern with mechanical properties, a comprehensive study of etching reagents and etching technique on Class 50 iron was made. This cast alloy is essentially free from dendritic cementite so that the etching reagents used for white irons are totally ineffective for the development of a macropattern.

Jurick made some suggestions in a paper on the primary etching of gray cast irons, and presented pictorial reproductions at 40 diameters which approximated to micro rather than macro details. He suggested the use of ammonium persulphate for bringing out graphite in dendritic structures and this reagent served as the foundation for the etching procedure subsequently developed by the present authors. The use of formic acid with concentrated sulphuric acid suggested by Jurick has no apparent effect in developing the structure of the irons studied.

Many other common combinations of etching solutions were also tried, but they gave results that did not appear to provide the slightest differentiation between the ill-defined structures of the various samples. It was not until the following etching procedure was developed that the authors were able to begin actual studies on the primary structure. A rough grinding operation through a series of belts, grit Nos. 80, 120, 240, and 320, was the only preparation necessary prior to etching in the following manner:

(1) 2.5 gm. (0.088 oz.) of ammonium persulphate in 100 cc. of water, specimen swabbed for 15 min.

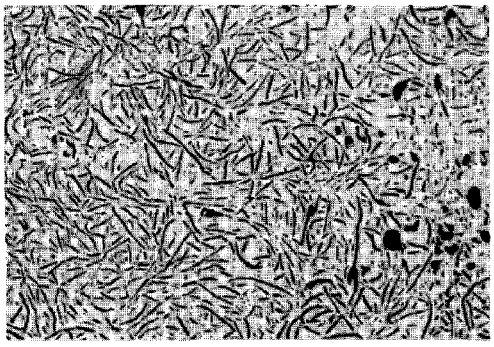


Fig. 16—Polished unetched gray iron section from a truck brake drum showing the coarse graphite flakes, under vertical illumination at 10 diameters.

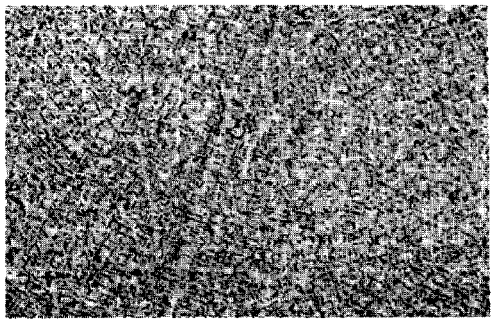


Fig. 17—A polished unetched gray iron section, showing the typical coarse "rosette" and "whorl" graphite formation, vertically illuminated at 10 diameters magnification.

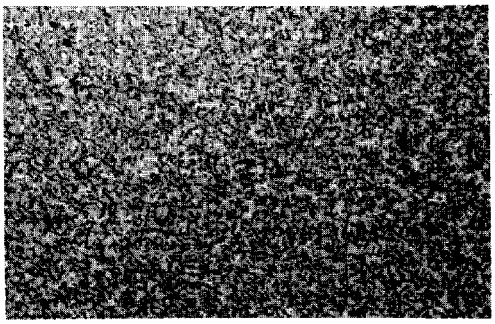


Fig. 18—A polished, unetched Class 50 iron section, indicating fine graphite distribution associated with dendritic pattern, under vertical illumination at 10 diameters magnification.

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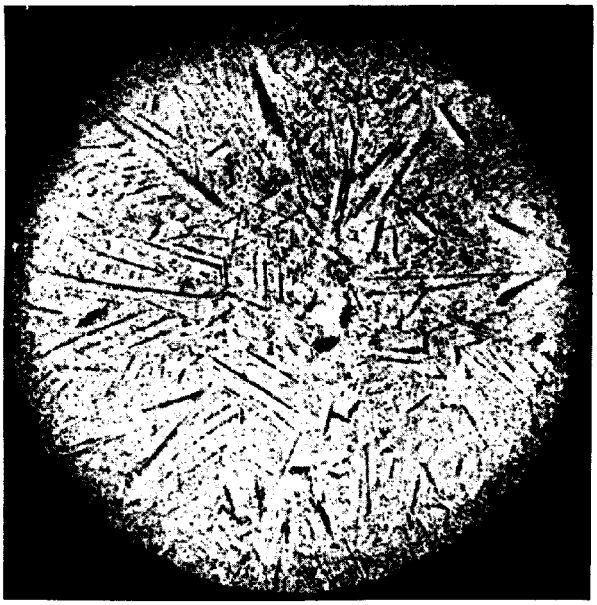


Fig. 19—The development of primary macrostructure in the Class 50 iron can be observed here. Using the new etching technique, the coarse, primary, dendritic structure may be noted. Enlargement was three diameters and the tensile strength of the piece is 50,000 lb. per sq. in.

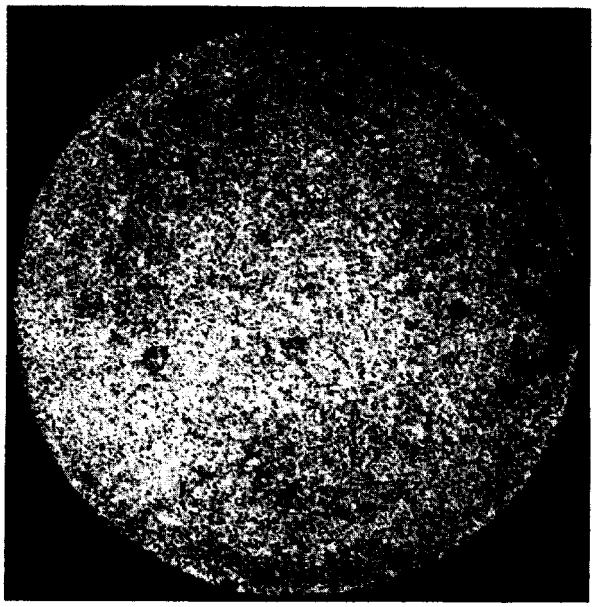


Fig. 20—Here may be noted the fine "macrostructureless" feature of Class 50 iron, free from any evidence of a dendritic pattern. Magnification three diameters, and tensile strength 40,200 lb. per sq. in.

(2) 1.5 gm. (0.0428 oz.) of potassium iodide added and specimen swabbed for 10 min. more.

(3) 1.5 gm. (0.0428 oz.) of mercuric chloride added and specimen again swabbed for 5 min.

(4) 15 cc. (0.9172 oz.) of sulphuric acid added and swabbing continued for 5 min.

The specimen was finally washed in water and dried with alcohol. No doubt some variations in the amounts of reagents added would not seriously modify the results obtained, but the amounts suggested appeared to provide optimum conditions. The results obtained from this procedure were very informative and a clear picture of the primary structure of the cast iron could readily be seen by a visual examination with the unaided eye.

The method of reproduction available has been discussed previously and the etched figures herein presented were recorded at three diameters with the Bausch and Lomb 72 mm. micro tessar using vertical illumination.

In Figs. 19 and 20 may be noted the etched structure of the transverse section of two Class 50 irons of identical chemical composition, but poured from the ladle at two different casting temperatures. The effect of pouring temperature is clearly profound.

Tests on sample bars

Four arbitration bars of Class 50 iron were cast from progressively lower pouring temperatures. Tensile test specimens were machined according to the A.S.T.M. standard specifications for gray iron, 0.505 in. diameter and 1 in. gage length. Transverse strength, deflection, and Brinell hardness data were also obtained. The results recorded on bars poured at 2700, 2530, 2470 and 2390 deg. F. are shown in Table I.

The transverse etched macrostructures of these four bars are illustrated at three diameters in Figs. 21, 22, 23, and 24. In the first three figures, the presence of primary

Figs. 21 to 24—Effect of pouring temperature on the etched macro pattern of Class 50 gray iron.

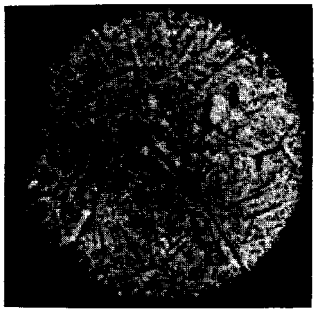


Fig. 21—Poured at 2700 deg. F.



Fig. 22—Poured at 2530 deg. F.

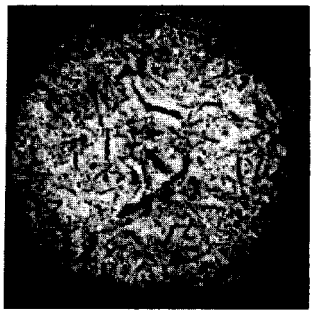


Fig. 23—Poured at 2470 deg. F.

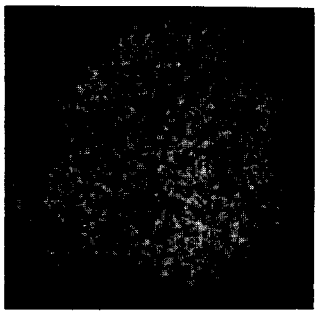


Fig. 24—Poured at 2390 deg. F.

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dendritic structure is clearly shown. However, it may be noted that the casting made from the highest pouring temperature studied, and exhibiting the highest tensile strength of the group, shows the persistence of dendrite formation to the peripheral parts of the bar.

In the case of the bar poured at the lowest temperature, 2390 deg. F. in Fig. 24, a fine "macro structureless" feature, free from any evidence of dendritic pattern, is obtained. With the bar poured at a slightly higher temperature, 2470 deg. F., it may be noted that the dendritic pattern is confined to the center portion of the arbitration bar.

It is of interest to point out that in cast materials the coarse dendritic patterns observed in Figs. 21 and 22 have usually been associated with low mechanical strength, whereas the fine uniform dendrite-free pattern obtained with low pouring temperature, shown in Fig. 24, has been associated with superior mechanical properties.

Reference to the tabulated data will reveal that in these castings, high casting temperature gives the most fully developed dendritic structure and also results in the optimum tensile property.

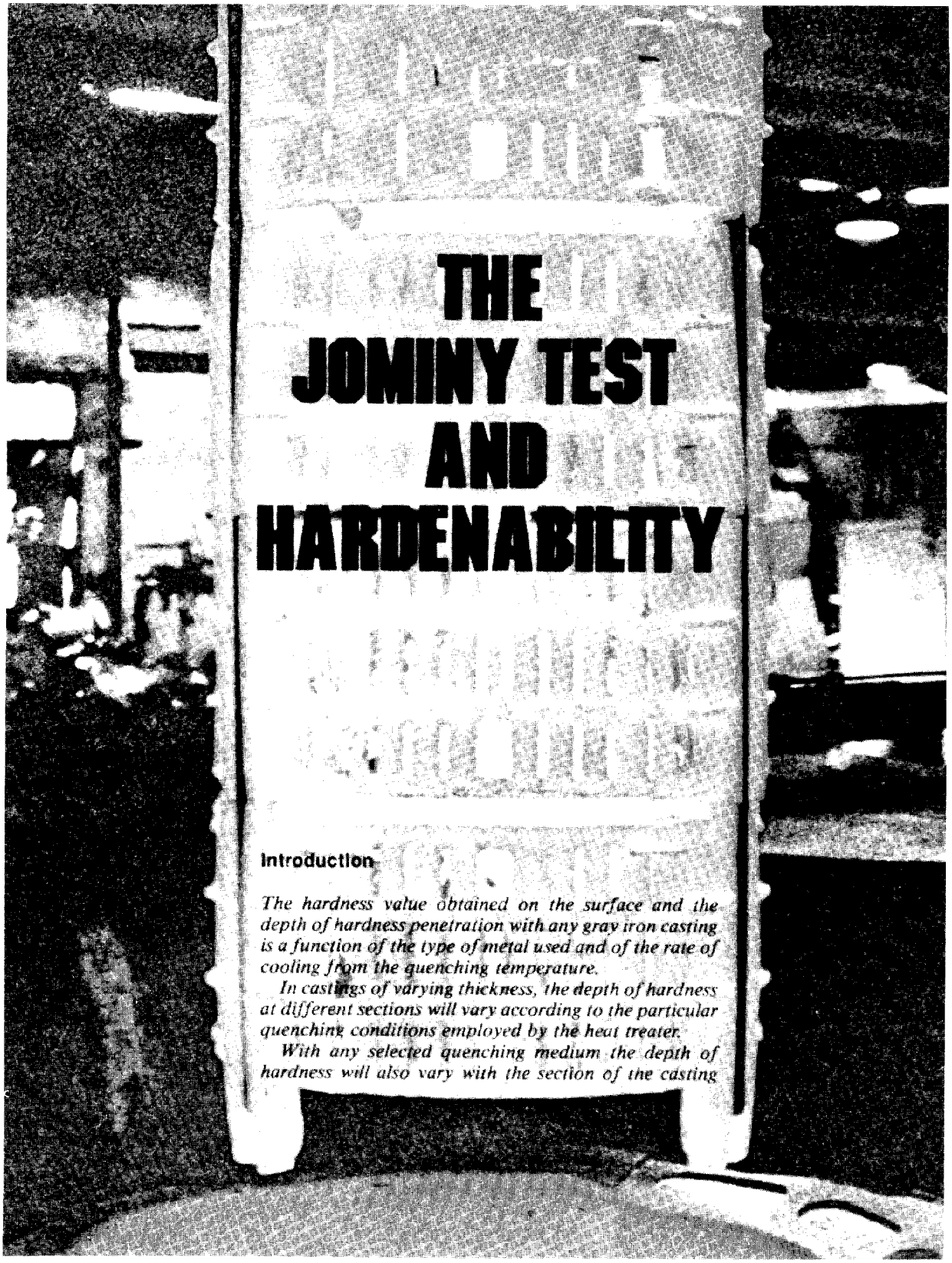
Referring back to Figs. 19 and 20, which were presented to indicate the development of a macro etching technique, it may be stated that a similar correlation was observed. In this series of arbitration test bars the casting made at the higher temperature and slowing marked dendrite formation, shown in Fig. 19, gave a tensile of 50,000 lb. per sq. in., whereas that cast at a lower temperature and entirely free from macro dendrites showed a tensile of 40,200 lb. per sq. in. A third cast in this series poured from the highest casting temperature had a tensile strength of 58,200 lb. per sq. in. The macro-picture has not been reproduced but the structure was in general similar to that shown in Fig. 19, except that the dendrites were slightly coarser.

Further investigations on the correlation between macrostructure and tensile strength are likely to reveal that the reason for the association of high tensile properties with coarse macro dendritic growth is dependent on the following concept: When conditions of casting are such as to lead to the formation of a coarse dendritic appearance on a macro etched section, these conditions promote the formation of very coarse austenitic dendrites during solidification. On cooling through the critical, the long dendrites transform to continuous pearlite colonies which confer strength to the casting. □

TABLE I
Results of Tests Made on 1.2-In. Diameter Arbitration Bars

Test	Pouring Temperatures			
	2700 Deg. F.	2530 Deg. F.	2470 Deg. F.	2390 Deg. F.
Tensile, lb. per sq. in.	49,100	47,300	42,600	39,400
Transverse strength, lb. per sq. in.	3,200 lb.	3,000 lb.	2,950 lb.	2,850 lb.
Deflection	0.430 in.	0.385 in.	0.360 in.	0.385 in.
Brinell hardness	217	217	217	217

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THE JOMINY TEST AND HARDENABILITY

Introduction

The hardness value obtained on the surface and the depth of hardness penetration with any gray iron casting is a function of the type of metal used and of the rate of cooling from the quenching temperature.

In castings of varying thickness, the depth of hardness at different sections will vary according to the particular quenching conditions employed by the heat treater.

With any selected quenching medium the depth of hardness will also vary with the section of the casting

Source: Casting Engineering/ Foundry World, Spring 1981

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since a thin section will cool relatively rapidly as compared with a heavy section similarly quenched.

One purpose of this article is to discuss the property characteristics of hardness and hardenability as determined by the Jominy end-quench test procedure which has become an established test method for determination of the reaction to hardening of ferrous metals on quenching in various media.

Of these two property characteristics the term hardness is most readily defined. It refers to the value for maximum hardness which may be obtained on the surface of the metal, but it provides no indication on the depth below the surface to which this maximum or absolute hardness may be found to penetrate.

The term hardenability serves as a measure of the depth of hardness and the Jominy test provides an index of this hardenability or hardness depth obtained on a given metal when quenched with a water spray under defined conditions.

A second purpose of this article serves to show the relation of the Jominy index of hardenability to commercial heat treating operations and to emphasize the importance of both mass effect and severity of quench in attempting to arrive at the depth of hardening of a casting from prior Jominy data.

Thus a die which has to be hardened is frequently cast with considerable variation in section. The following discussion will demonstrate that depth of hardening is a variable and that the light section may harden to a considerable depth while the heavy section may exhibit shallow hardening. It is shown, however, that the effect of mass on this variation in depth of hardening varies with the severity of quench (water versus oil) and that the less severe the quench the greater effect will mass variation have on the variation of the depth of hardening at different sections of the casting and possibly on the hardness value of the outer zone of the casting.

Since there is an unlimited number of shapes to be hardened, it is impossible to draw up a correlation between these many shapes, the Jominy end-quench index, and the severity of quench used. However, the direct relation of the Jominy test in relation to commercial hardening of round sections quenched in different media has been clearly indicated.

The use of the Jominy test as an index of quality has also been emphasized.

Several charts are presented to demonstrate:

(1) The hardenability of several type Class 50 metal castings.
(2) The effect of alloy additions on hardenability characteristics.
(3) The hardenability of types Class 45, 40 and 30.

The Jominy Test and Hardenability

When a section of cast iron is hardened by quenching, the extent or depth to which it is hardened is governed by:

(1) The hardening capacity of the material to be quenched, and
(2) The rate at which the piece cools, which is governed by both the size of the casting and by the type of quenching media—oil or water.

The potential hardening capacity depends on the type of metal used and on its chemical composition particularly with respect to silicon and manganese content. This potential hardening capacity is known as the "hardenability" of a metal.

On account of the increasing importance of heat treating engineering castings for both improvement in tensile and impact properties as well as for wear resistant purposes, this "hardenability" characteristic and the factors which affect its magnitude must receive increasing attention.

Several methods have been used to measure this property of hardenability. They include the Jominy end-quench test, the cone test, the Shepherd fracture test and the etch test on polished round bars of varying diameter.

The purpose of all these tests is to determine to what depth the metal will fully harden when quenched in a given medium independent of the absolute or surface hardness value attained.

The maximum or absolute hardness value which can be produced on the surface of a simple iron-carbon alloy or of one containing any of several alloying elements is dependent wholly on carbon content. This is clearly indicated in Figure 1, which shows the maximum hardness in terms of Rockwell "C" as a function of carbon content.

The data were obtained from plain carbon and from alloy S.A.E. steels and it may be observed that the hardness as defined by resistance to deformation of a penetrator is independent of the alloy content of the steel.

The hardness does depend, however, on the carbon content of the metal when below about 0.6 percent carbon but with increasing carbon content beyond this amount, full hardness can be obtained independent of carbon analysis.

The matrix of a gray iron engineering casting should always contain not less than about 0.6 percent carbon so that we are not concerned with absolute hardness, and as we have observed, alloying elements will not affect this absolute hardness to any appreciable degree.

It may be thought that all engineering grades of cast iron should, therefore, give the same Brinell hardness when fully quenched, but it must be remembered that the flake graphite forms an appreciable volume of the structure. As the amount of graphite is progressively increased from type Class 50 to Class 30 the full hardness as measured by resistance to penetration of a Brinell ball or other penetrator is progressively decreased since graphite has little resistance to this type of test.

Hardenability, however, is a very different property

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and is dependent on such factors as:

(1) Grain size.
(2) Alloy content.
(3) Rate of cooling as depending on both the quenching medium and the degree of agitation.

Whatever these factors may be, it will be readily recognized that they may be classified into two groups which are concerned with:

A. The inherent hardenability of the metal.
B. The quenching power of the coolant.

Inherent Hardenability

It is well known that when a gray iron casting is heated just above the critical range, the pearlite changes to austenite. If the temperature is further raised, some of the flake graphite (and any free carbide which may be present) also tends to pass into solution and thus raise the carbon content of the austenite.

Thus at all temperatures above the critical range, short of melting, a simple structure is obtained which consists of grains of austenite containing some 0.8 percent or more of dissolved carbon as the matrix, with excess free flake graphite or cementite in quantity and distribution essentially similar to that observed in the casting at room temperature.

When the matrix structure obtained at temperatures above the critical is cooled, it changes either to pearlite or martensite dependent on the rate of cooling. The cooling rate which must be exceeded to prevent the formation of pearlite and hence to assure austenitic decomposition to hard martensite is called the "critical cooling velocity".

Accordingly, any portion of a casting which is cooled at a rate exceeding this "critical cooling velocity" will show a full martensitic hardening from a transformation occurring at 300 to 400°F. When the cooling rate approximates that of the critical cooling velocity, a mixture of martensite and pearlite will be obtained.

If a casting is to be fully hardened to the center this can clearly be accomplished only if the cooling rate at the center of the casting exceeds the critical cooling velocity of the type metal poured.

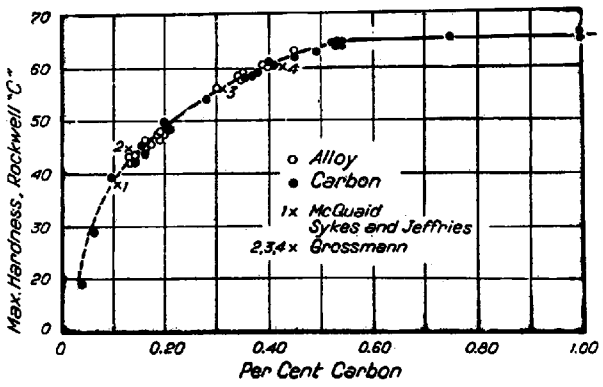


Figure 1—Showing effect of carbon content on maximum surface hardness which can be obtained on quenching. The data also indicates that alloy additions do not materially change the maximum hardness value.

As the quenching medium is changed from oil to water, we increase the severity of quench but with heavy castings a severe water quench may be insufficient to cause a rate of cooling at the center of the casting, exceeding the critical cooling velocity. In such instances the casting will have a core showing a pearlite structure after quenching.

The factors which are chiefly responsible for changes in critical cooling velocity include:

1. Type or class of casting in.
2. Time and temperature of holding above the critical range.
3. Alloy content—the most important factor.

It will be noted that the size or thickness of the casting has no bearing on "critical cooling velocity". The critical cooling velocity, however, is the factor which determines either:

a. How large a casting can be fully hardened to the center or
b. What depth of hardening will be obtained on castings which are too thick to harden fully to the center.

Brief reference should also be directed to the method of quenching and the quenching medium employed.

Assume that still warm water or that any mineral oil be used instead of rapid agitation in cold water then the rate of heat removal from the metal will be markedly decreased and the rate of actual cooling at a depth of say 1/8 inch from the surface of the part will be materially decreased. Thus the hardened case of a given casting may be reduced from 1/8 inch to 1/16 inch since the slower rate of cooling has assured that the critical cooling velocity has now been exceeded only to a depth of 1/16 inch in warm water or mineral oil as compared with 1/8 inch in agitated cold water. The inherent hardenability has not been changed but the *depth* of hardness obtained for any given iron is clearly a function of quenching medium or degree of agitation.

From the above discussion, it is manifest that if we desire to examine the inherent hardenability of different gray iron castings we must standardize on some specific and reproducible method of quenching. This has been accomplished in the following standard specification which has been developed for testing steel.

End-Quench Test for Hardenability

This is the Jominy end-quench test in which a 1 inch diameter cylinder of the metal is heated to above the critical and then quenched on one end. The nature of the test is clearly indicated in Figure 2 which shows a test sample in the process of end quenching.

After treatment, the side of the specimen is ground flat to about 0.015 inch depth and hardness readings in terms of Rockwell "C" made on the test specimen in steps of 1/16 inch. The plot of hardness as a function of

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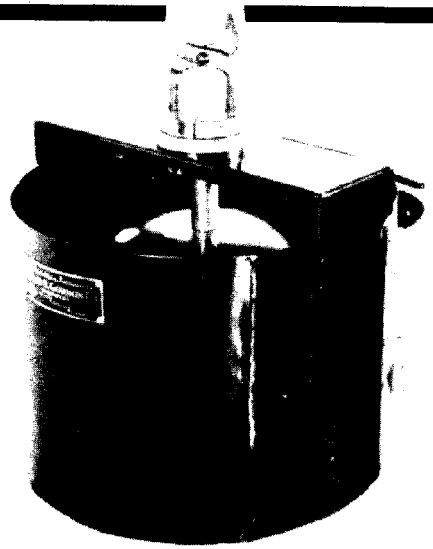


Figure 2—Photography showing the Jominy End-Quench being given to a standard 1 inch diameter bar.

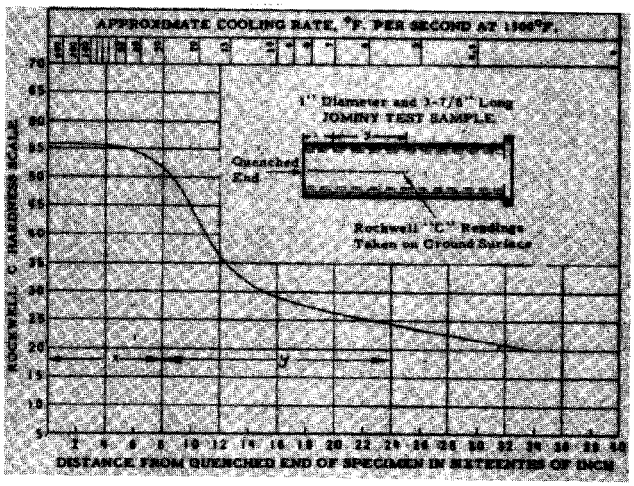


Figure 3—Plot of hardness data with typical end-quench hardness curve obtained on type "GA" Meehanite metal representing a class 50 gray iron.

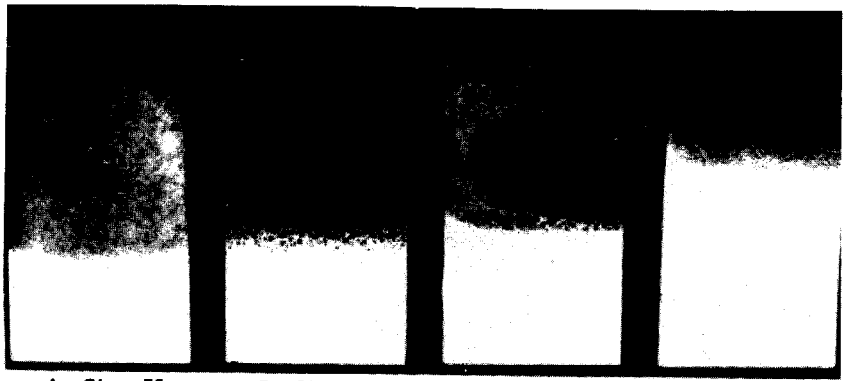
the distance from the quenched end provides information on:

(1) Maximum hardness obtained.
(2) Depth of hardenability.

It will be noted that such data are dependent entirely on the characteristics of the test bar. The quenching procedure is constant and approximates maximum severity of quench. This merely implies that the end surface of the bar is immediately reduced to the temperature of the quench medium and that the cooling rate at increasing distances from the quenched-end is progressively but constantly lower. The constant progressively lower rate is dependent on the heat conductivity or diffusivity of the metal which is assumed to be constant for ferrous base materials.

A typical curve obtained from data taken from a standard type "GA" Meehanite metal casting is reproduced in Figure 3. It will be noted that full hardness (Rockwell "C" 55 in this case) extends from the quenched-end of the bar for a distance of 6/32 inches. The approximate cooling rate at this location is 45°F. per second. Rockwell "C" 45 is often accepted as a measure of the depth of hardening. In this case the casting may be said to fully harden to a depth of 10/32 inches and that the critical cooling velocity of the iron approximates a cooling rate of 20°F. per second. Any section of the casting cooled at a slower rate than this will fail to harden.

It is of interest to examine the etched macrostructure of Meehanite metal test bars which have been end-quenched and then sectioned longitudinally through the center of the bar. Figure 4 illustrates four such sections from tests conducted on test specimens prepared from the standard 1.2 inch diameter Meehanite metal arbitration bars. The white areas have a hard martensite structure and the dark areas are pearlitic. The relative shallow hardening of the first three castings as compared with the deep hardenability of the fourth casting is readily observed. The transition zone is most clearly revealed in



A—Class 50	B—Class 50	C—Class 50	D—Class 50
with 3.09 C.	with 3.07 C	with 3.15 C.	with 2.99 C
1.22 Si.	1.26 Si.	1.21 Si.	0.93 Si.
0.84 Mn.	0.86 Mn.	0.85 Mn.	1.21 Mn.

Figure 4—Illustrates the appearance of polished etched longitudinal sections cut from Jominy Bars. The hard white martensitic areas extend down from the quenched end to the pearlitic unhardened regions shown darking in the picture.

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the second photograph from the left.

With a simple plain carbon steel examination of the hardness distance plot will usually show two essentially horizontal sections; the first at a level of about 60 Rockwell "C", the second at a low level of perhaps 20 Rockwell "C". These two sections of the curve are joined by a steep section showing a rapid fall in hardness from the martensitic "C" 60 to the softer pearlite structure where the quenching rate has failed to exceed the critical cooling velocity.

In common engineering steels and in Meehanite metal, and most engineering gray irons the alloy content for elements such as silicon and manganese prevent this sharp transition. The gradual transition can be accounted for by the presence of a mixed structure or transition zone consisting of varying amounts of martensite and fine pearlite. It may be considered as comparable to the mottled transition zone found in white irons having a gray structure in the core of the casting.

For this reason, recourse has been made to an "index of hardenability" by which the hardenability is designated by a code indicating the distance from the quenched end of a specimen within which the designated hardness is obtainable. Thus J_{45} = 7 listed in a specification means that a minimum requirement would be a Rockwell hardness of "C" 45 at a distance of 7/16 inches from the quenched end. This idea has been discussed and illustrated in Figure 3.

Significance of Hardenability Tests

The Jominy end-quench test is now used as a purchase specification for steels which are to be heat treated and hardened for engineering applications. Since it is not practicable to manufacture steel or any engineering metal to an exact chemical specification neither is it practicable to manufacture steels or Meehanite or gray iron castings giving a definite Jominy hardenability curve.

Accordingly, resort has been made to specifications in terms of "hardenability band". This is comparable to tolerances permissible in chemical analysis.

Such variability in reaction to hardening will always be common to Meehanite and gray iron castings so that Jominy hardenability bands will be necessary to indicate the reaction to quenching of different engineering types of metal.

Since recourse is never made to the end-quench procedure for hardening parts to be used in service, it becomes necessary to examine the significance of the Jominy data in terms of normal quenching operations.

On account of the variety of shapes of fabricated or cast parts which are quench hardened, it is impracticable to attempt to survey and generalize the many conditions which may be encountered.

It has become customary, however, to consider the relation between the extent of Jominy end-quench hardenability and the reaction of round bars of varying diameter to depth of hardening (hardenability) in various quenching media.

In these comments, we propose to briefly survey two specific aspects of the subject.

1. The diameter of bar which results in complete hardening and hence in the absence of a soft core.
2. The effect of different quenching media and their severity of quench.

It is common knowledge that a greater depth of hardening will be obtained with any given iron-carbon alloy, the less the diameter of the section to be quenched. Similarly, for any given diameter bar, the depth of hardening will be greater, the greater the severity of the quench treatment. A comparison of the quenching power of oil and water furnishes an excellent example for the latter statement.

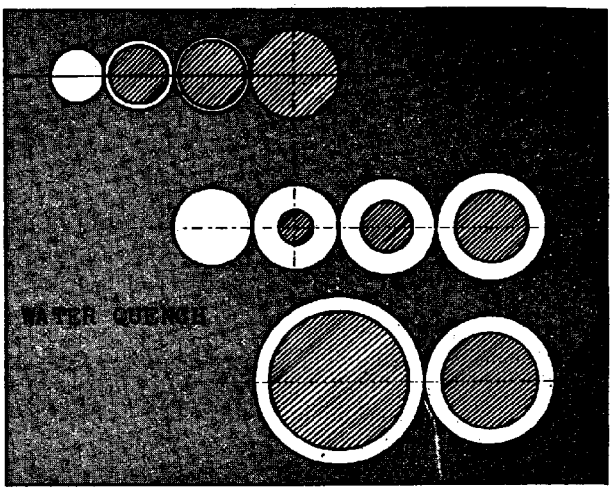


Figure 5—Diagrammatic representation of the reaction to quenching in water and in oil of round bars. The white areas indicate martensite or full hardening. The shaded areas indicate pearlitic regions which failed to respond to hardening.

Both statements are simply and fully epitomized in Figure 5. Thus the diagrams illustrate:

1. Depth of hardening is greater, the more severe the quench. Use of water instead of oil.
2. Using a constant quench, depth of hardening becomes less as the bar size is increased.
3. With the milder (oil) quench the diameter of the unhardened core increases more rapidly with increase in bar diameter than is the case with the more severe (water) quench.

Whatever the quenching medium selected, we arrive at a bar diameter in which the soft core has disappeared and the bar is just fully hardened to the center, by a process of progressively decreasing the bar size which is to be quenched. This bar diameter has been termed the "critical size" for the given metal for the severity of quench under consideration.

Now this "critical size" will naturally vary for different quenching media (see Figure 5) and the maximum size will be obtained only with maximum severity of quench. Such a diameter of bar is then defined as the "ideal critical size".

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The dimensional relationship of the hardenability characteristics shown in Figure 5 are given in Figure 6. The unhardened diameter (Du) will always be less than the bar diameter unless complete absence of hardened periphery obtains as is the case of the 2 inch diameter bar quenched in oil.

A family of curves like AB and CD can be drawn for all varieties of quenching media for the particular steel

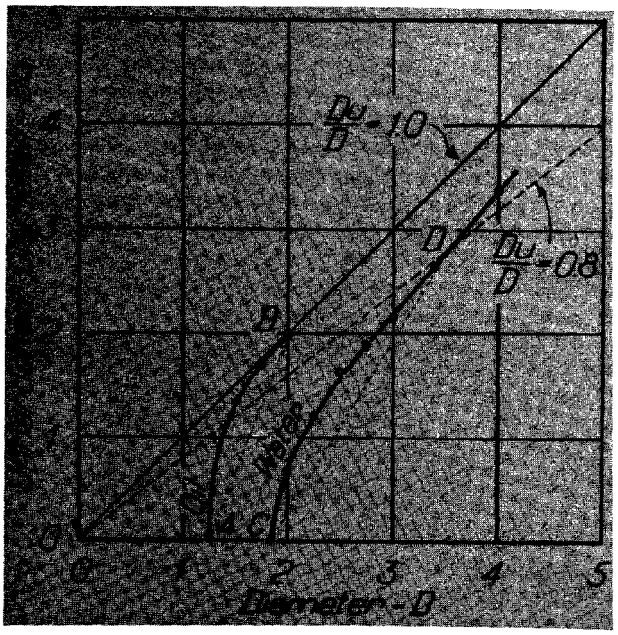


Figure 6—Characteristic curves depicted in Figure 5. The graph gives the specific relation of bar diameter (D) to unhardened core diameter (Du). A bar less than 1¼ inch diameter will fully harden to the center when quenched in oil, whereas a bar less than 1¾ inch diameter will fully harden when quenched in water.

or iron represented in Figure 5. Similarly a family of curves can be prepared for metals of other compositions and reactions to hardenability.

The sharp transition from hard martensite (white) to pearlite (shaded) shown in Figure 5 is not obtained on quenching Meehanite castings since a transition zone usually accompanies the quenching operation. This transition zone of mixed structures is indicated by the gradual drop in hardness shown in Figure 3 between approximately 8/32 and 12/32 distance on the Jominy end-quenched bar.

Finally we should consider the relation of the Jominy end-quench data to the round bar hardenability as discussed in the preceding paragraph.

The most usual method of inter-relating these two quenching methods is to indicate the bar diameter which will just harden to the center as a function of hardenability found by the Jominy method. Any curve which expresses this relationship must be related also to some specific intensity or severity of quench.

In Figure 7, three different curves have been included. Those for water and for a regular mineral quenching oil refer to vigorous agitation of the test bar in the quenching medium. The curve for "still air" approximates the results obtained from a normalizing treatment with deep hardening materials.

The significance of this figure can best be demonstrated by taking a specific example. Thus assume that a 1 inch diameter Jominy bar be cut from a given casting and that on test it is found that the extent of Jominy hardening is 8/16 inches. We wish to find what maximum thickness of casting will fully harden when vigorously quenched in oil. From Figure 7, it can be noted that 8/16″ or 1/2″ on the "oil" curve corresponds to approximately 1¼ inches diameter for the bar with the equivalent center cooling rate. Thus a bar cast to 1¼ inch and vigorously quenched in oil is the maximum size which will fully harden to the center.

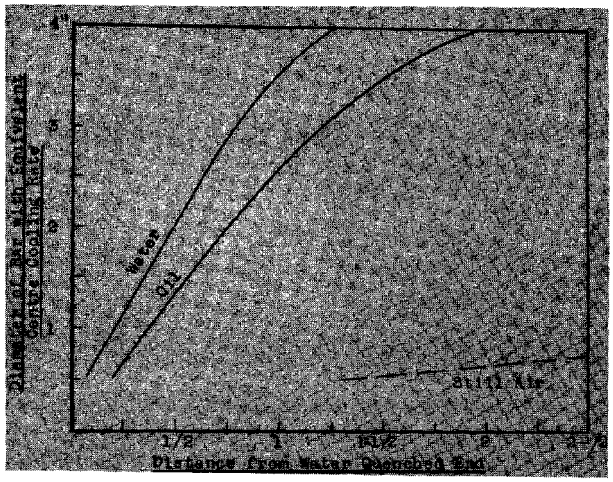


Figure 7—Relation between Jominy hardenability and diameter of bar which will fully harden. Thus, a 1¼ inch diameter round bar quenched in oil should fully harden to the center with material which hardens to ½ inch in the Jominy end-quench test.

Summary On The Use Of Jominy Test

The use and meaning of the Jominy end-quench test may be summarized as follows:

1. The present investigation relates to "hardenability", the means of measuring this property and the significance of the measurement registered.
2. The investigation is concerned with martensitic hardening whether attention is directed to hardness or hardenability.
3. The hardness which can be induced by any quenching operation refers only to the maximum hardness value which can be obtained on the surface of the casting. This maximum hardness can be obtained only when the "critical cooling velocity" of the casting is exceeded at the surface. It fails to provide any index of the depth below the surface to which the casting will be hardened.

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4. The Hardenability provides the index relative to depth of hardening which can be induced by any quenching operation. This hardenability depends both on carbon content and on alloy additions. Since alloy additions lower the critical cooling velocity, an increase in Mn., Cr., Ni., Va., Mo., etc. increases the depth of hardened case of the casting for any constant quenching operation. The potency of these elements, however, differs widely. Similarly for castings with or without alloy addition the hardenability (depth of hard case) is increased with increase in severity of quench operation (eg. use of water instead of oil).

5. The most common method for measuring the hardenability of an iron-carbon base material is termed the Jominy end-quench test. This necessitates the preparation of a 1 inch diameter bar about 4 inches long. The bar is heated to 1575 °F. for Meehanite metal, rapidly placed in a fixture (Figure 2) and cold water sprayed on one end of the bar for a minimum period of 10 minutes. After grinding a flat on the side of bar, Rockwell "C" readings are taken every 1/16 inch from the quenched end. The maximum distance transversed for full hardness (in 1/16 inch units) is termed the Jominy Hardenability index.

   If a bar prepared from one casting fully hardens to 3/16 inch while a bar prepared from a second casting fully hardens to 6/16 inches in the Jominy test, then the second bar has double or twice the hardenability index of the first bar.

6. The importance of this Jominy hardenability index lies wholly in its translation in terms of hardenability of an engineering casting which may be simple in design or of irregular shape. In the Jominy test the severity of quench approaches 100 percent or the maximum severity of quench obtainable. Such a quench is rarely used in industry.

   It is, therefore, necessary to correlate the results of the Jominy test with *each specific* casting design and with the mode of quenching operation. This latter includes quenching medium and method (still or agitated) of quench employed.

   The effect of severity of quench needs no further emphasis but the effect of design (size and shape) requires further comment.

   Assume that the Jominy end-quench figure is 8/16 for a type metal. Now assume that two castings, 1 inch and 4 inches thick respectively, are quenched in cold water in order to obtain a similar high intensity of quench. The depth of hardened case in the 4 inch casting bar will be much less than in the 1 inch diameter bar. This has been typified in Figure 5 and dimensionally considered in Figure 6 for one given metal.

   Thus from the latter figure it may be noted that a bar of this steel up to a maximum of 1¾ inch will have zero unhardened core when water quenched. Obviously, a 1 inch bar will also fully harden to the center. On the other hand, a 4 inch diameter bar of this particular material will have a 3 inch unhardened core after water quenching. The depth of hardened case will therefore be ½ inch.

   Accordingly, it is necessary for the individual foundries to establish the relation between the Jominy end-quench index and the depth of hardening which can be obtained for any group of castings of similar dimensions.

   The Jominy test is then used as a control method for checking and maintaining a constant reaction to hardenability of the particular type of metal under consideration.

7. Simple relationships have been established between the Jominy index and full hardening (hardening to the center) of various sized bar diameters, when quenched in different media but under constant conditions for the same quenching medium. This relationship, however, provides no data for determining specifically the depth of surface hardening which will be obtained in bars of greater diameter similarly quenched. The approximate relationships for full hardening of one particular material in water, oil and still air have been reproduced in Figure 7. For other quenching media other curves can be constructed to form what may be termed a family of curves. Similarly for each different material there is a different but specific new family of curves.

   From an examination of the curves given in Figure 7, we may state that if a bar of 1 inch diameter is to harden fully to the center when quenched in oil then the Jominy hardenability index must be about 8/16 to 9/16. On the other hand, if the 1 inch diameter casting is to be quenched in water, full hardening can be obtained if the Jominy index is not less than 4/16.

   The Jominy hardenability index clearly indicates whether any casting of one engineering type of metal will harden more deeply than a similar casting of a different engineering type of metal. The degree of difference in hardness depth must be experimentally determined and recorded by the heat treating department.

   By use of such correlation, the Jominy index is of value in conventional quench hardening and in flame and induction surface hardening treatments. Without this correlation, however, the Jominy end-quench index is still useful as a measure of the inherent hardenability of any type of metal.

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Jominy End-Quench Characteristics of Certain Cast Irons

This report provides a survey of the Jominy end-quench characteristics of several types of cast iron. The data furnish information on:

1. Maximum hardness and
2. Hardenability—depth of hardness which can be obtained on spray water quenching.

The relative difference in hardenability indicated on the curves also serves as an index of the relative depth of hardening under less drastic quenching conditions such as are obtained with the use of ordinary quenching oil.

The effects of single element additions on type "GA" base Meehanite metal castings including molybdenum, manganese, copper, cobalt, aluminum and chromium are recorded as well as the effects of some of the elements in combination.

Chart I-Typical type GA50 Meehanite metal cast in 1.2 inch diameter bar.

	ANALYSIS		
X	3.12 C.	1.49 Si.	0.91 Mn.
Y	3.05 C.	1.48 Si.	0.94 Mn.
Z	3.02 C.	1.57 Si.	0.94 Mn.

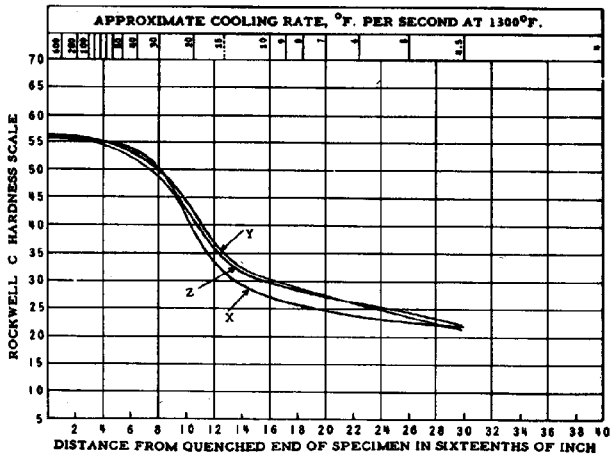


Chart II—Effect of Manganese on Hardenability of Typical Meehanite Metal Type "GA" with 0.79 Manganese Content

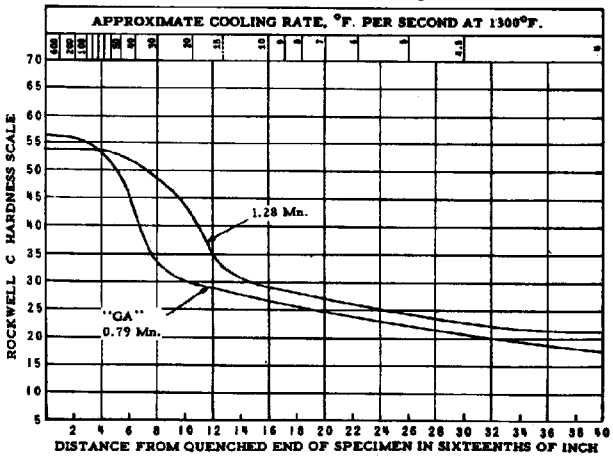


Chart III—Effect of Molybdenum on Hardenability of Typical Meehanite Metal Type "GA" with 0.79 Manganese Content

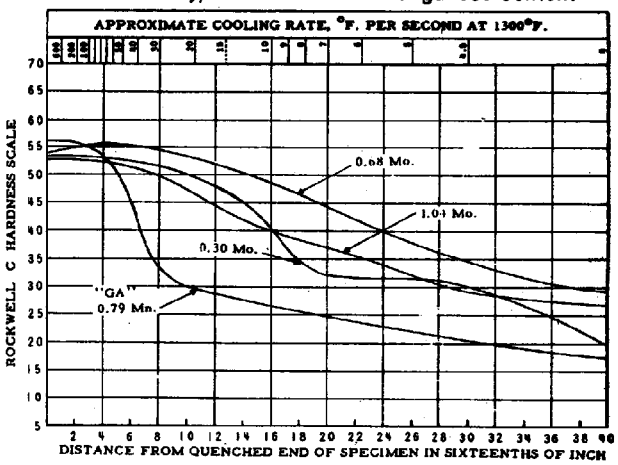


Chart IV—Effect of Molybdenum on Hardenability

	ANALYSIS			
A	2.94 C.	1.15 Si.	1.02 Mn.	—
B	2.94 C.	1.15 Si.	1.02 Mn.	0.34 Mo.
C	2.94 C.	1.15 Si.	1.02 Mn.	0.62 Mo.

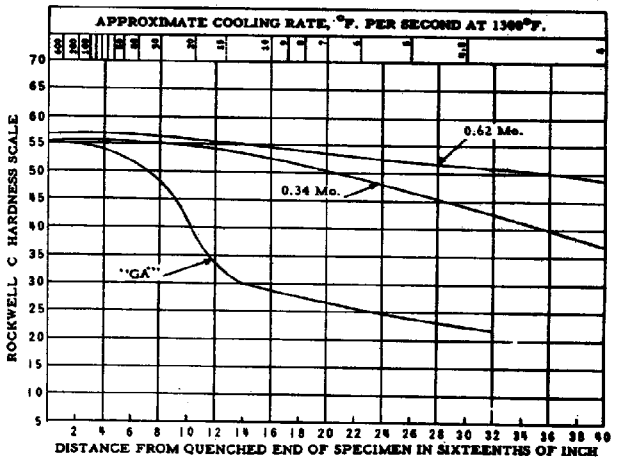
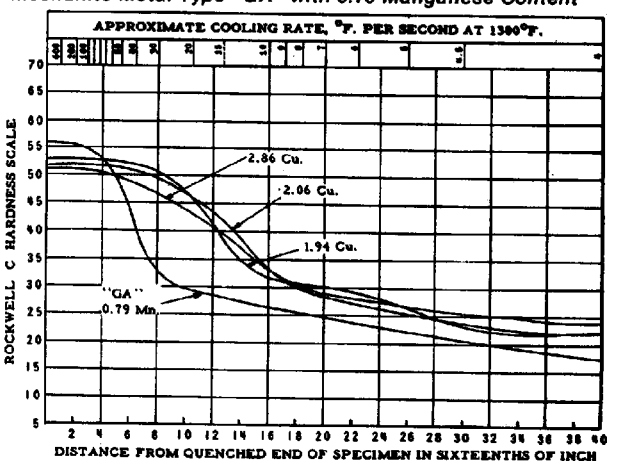


Chart V—Effect of Copper on Hardenability of Typical Meehanite Metal Type "GA" with 0.79 Manganese Content



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Chart VI—Effect of Molybdenum in Combination with Copper or Nickel on Hardenability of Meehanite Metal Type "GA"

ANALYSIS	A	2.94 C.	1.15 Si.	1.02 Mn.
	B	2.94 C.	0.34 Mo.	
	D	2.94 C.	0.32 Mo.	1.35 Cu.
	E	2.94 C.	0.15 Mo.	1.27 Ni.

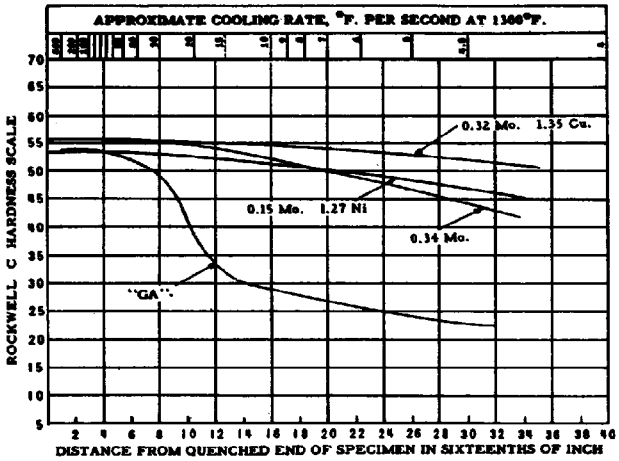


Chart VII—The Effect of 1/2 Percent Molybdenum in Combination with Chrome-Copper on Hardenability of Meehanite Metal Type "GA"

ANALYSIS	13.08 C.	1.28 Si.	0.68 Mn.	1.55 Cu.	0.7 Cr.	—
	23.08 C.	1.28 Si.	0.68 Mn.	1.65 Cu.	0.3 Cr.	0.5 Mo.
	32.82 C.	1.23 Si.	0.67 Mn.	1.61 Cu.	0.3 Cr.	0.47 Mo.

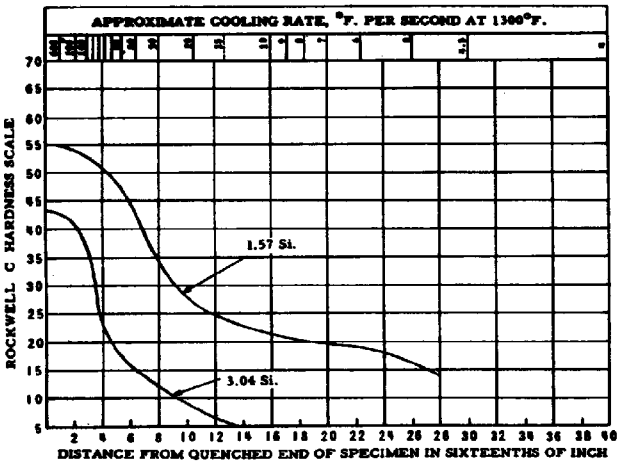


Chart VIII—Effect of Increasing Silicon Content on Hardenability of Meehanite Metal Type "GA"

ANALYSIS	CBCC	3.07 C.	1.57 Si.	0.48 Mn.	0.53 Cu.
	CB3	3.07 C.	3.04 Si.	0.42 Mn.	0.58 Cu.

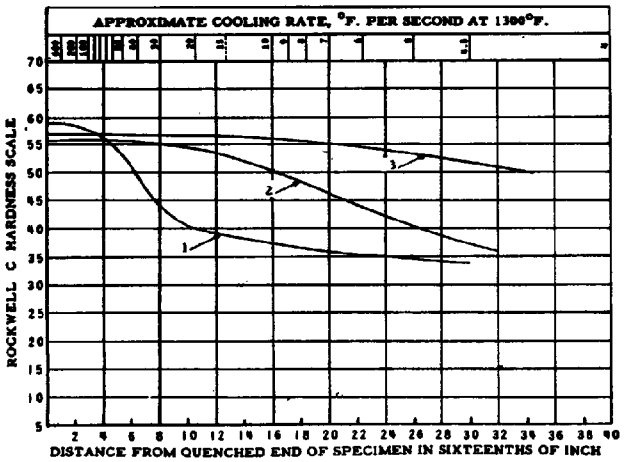


Chart IX—The Hardenability Characteristics of Type "GE" Meehanite Metal

ANALYSIS	#1	3.16 C.	1.74 Si.	0.86 Mn.
	#2	3.26 C.	1.77 Si.	0.85 Mn.
	#3	3.15 C.	1.84 Si.	0.89 Mn.

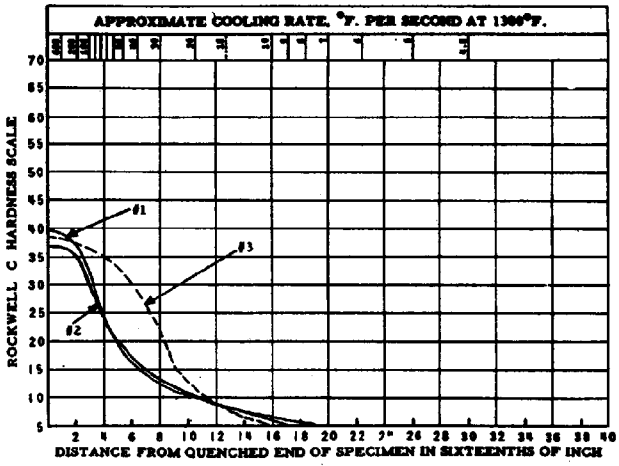


Chart X—The Hardenability Characteristics of Meehanite Metal Type "GC"

ANALYSIS	#1-2	3.27 C.	2.39 Si.	0.63 Mn
	#3	3.31 C.	2.35 Si.	0.68 Mn.

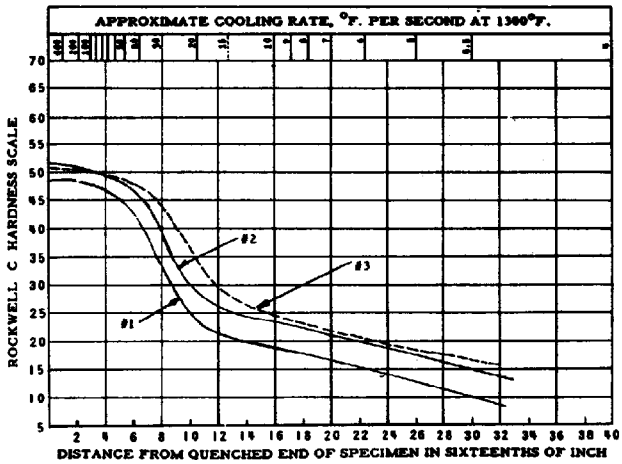
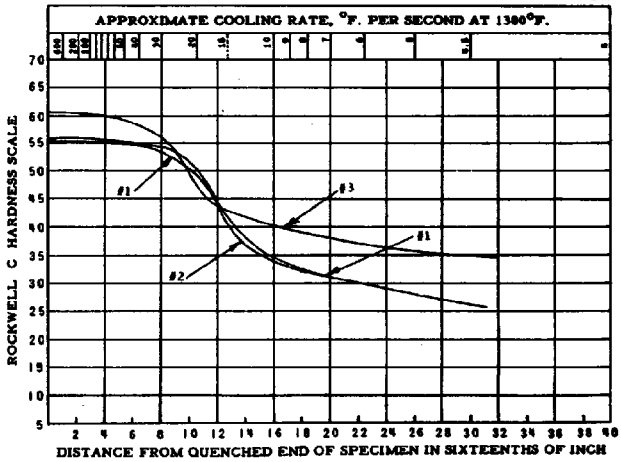


Chart XI—Effect of Chromium Additions on the Hardenability Meehanite Metal Type "GA"

ANALYSIS	#1	3.03 C.	1.18 Si.	1.11 Mn. 0.12 Cr.
	#2	3.05 C.	1.20 Si.	1.09 Mn. 0.24 Cr.
	#3	3.02 C.	1.19 Si.	1.09 Mn. 0.44 Cr.



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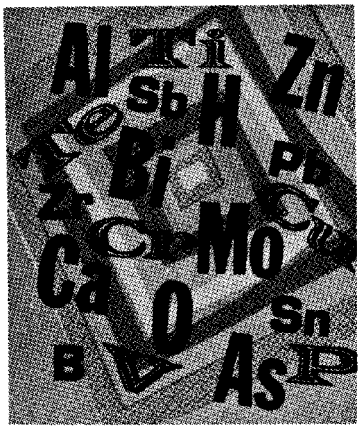
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Tramp Elements in Cast Iron

Minor elements in iron castings may have consequences far out of proportion to their amounts. Constant vigilance is needed by the foundryman, especially today when scrap contamination is often encountered. This article summarizes the effects of tramp elements on various members of the cast iron family of alloys.

By CARL R. LOPER JR. / Professor of Metallurgical Engineering, University of Wisconsin, Madison, Wis.

DURING THE processing of cast irons, certain elements may accumulate unintentionally to levels where they exhibit a noticeable effect on microstructure and properties of castings. When this effect influences the solidification process and/or the development of the matrix structure detrimentally, these elements are referred to as tramp elements (also termed subversive or residual elements). Most commonly their influence is observed at very small amounts; the total of all tramp elements in a given cast iron may be less than 0.1%.



Sources—Tramp elements originate from various sources—iron ore, steel or cast scrap, ferroalloys, ladle or furnace additions, coke, gases in melting, fluxes, etc. In general, control of tramp elements entails control and proper selection of all raw materials (and reliability of the suppliers) and a dependably reproducible process of melting and handling cast irons in the foundry.

Pig Irons—In the United States and Canada, we are fortunate in that iron ores, used to produce pig irons and virgin steel, are relatively free of contaminating tramp ele-

RESIDUAL ELEMENTS IN GRAY IRON[1]

ALUMINUM—*Principal Sources:* Ferroalloys and also nonferrous materials entrapped with cast scrap or steel. *Typical Amounts Observed:* .01-.05%. *Significant Effects:* Strong graphitizer, promotes graphite, greatly reduces chill tendency. Literature notes that .010-.10% may cause pinholes; effect is more severe as Ti content of iron is increased. Effects may differ in the base iron or ladle additions.

ANTIMONY—*Principal Sources:* Tramp element in cast scrap or steel, vitreous enamel on cast irons or steel; may be included with copper additions. *Typical Amounts Observed:* .005%. *Significant Effects:* Refines graphite somewhat, but said to produce abnormal structures. About .03-.06% required for fully pearlitic matrix. No apparent effect on chill. Even with all Type D grahite, ferrite is eliminated with .2% Sb.

ARSENIC—*Principal Sources:* Tramp element in cast scrap or steel; may be included with copper additions. *Typical Amounts Observed:* .01%. *Significant Effects:* No significant effect on graphite or chill. About .10% required for fully pearlitic matrix.

BISMUTH—*Principal Sources:* Usually a ladle addition. *Typical Amounts Observed:* .01%. *Significant Effects:* Increases chill substantially. Promotes Type D and mesh graphite. Effects reduced by inoculation and cerium treatment. Decreases eutectic cell count.

BORON—*Principal Sources:* Intentional addition. Present in pig iron, enameled scrap. *Typical Amounts Observed:* .01-.05%. *Significant Effects:* Promotes Type D graphite. Increases cell size. Promotes austenite dendrites.

CALCIUM—*Principal Sources:* Present in ferroalloys. *Typical Amounts Observed:* .04%. *Significant Effects:* Increases cell count. Larger amount increases chill depth. Increases inoculating effect of late Fe-Si additions.

CHROMIUM—*Principal Sources:* Alloying elements in cast scrap or steel. Also used as a ladle addition. *Typical Amounts Observed:* .03-.04%. *Significant Effects:* Greatly affects chill. Promotes pearlite in matrix, but not without tendency to promote primary carbides.

COPPER—*Principal Sources:* Nonferrous material in cast scrap or steel. Also used as a ladle addition. *Typical Amounts Observed:* .01-.90%. *Significant Effects:* Used primarily as means of insuring pearlitic matrix. Copper may be source of other residual elements.

HYDROGEN—*Principal Sources:* Exposure to gases in melting and prior to solidification. Wet furnaces and ladle refractories. *Typical Amounts Observed:* .0002-.0015%. *Significant Effects:* Said to promote increased chill. Also associated with pinhole formation.

LEAD—*Principal Sources:* Nonferrous material entrapped in steel or cast scrap. Also may enter with fluorspar, copper additions, or terne plate in steels. *Typical Amounts Observed:* .003%. *Significant Effects:* Chill increased substantially. Promotes Type D and mesh graphite. Effects somewhat eliminated by inoculation and cerium treatment. In combination with water vapor promotes Widmanstaetten graphite.

MOLYBDENUM—*Principal Sources:* As an alloy in steel or cast scrap. *Typical Amounts Observed:* .01-1.50%. *Significant Effects:* Strong pearlite former, increases hardenability. Mild carbide-forming tendency.

NITROGEN—*Principal Sources:* Exposure to gas in melting, particularly in arc melting. Can be introduced through high N core binders. *Typical Amounts Observed:* .002-.008%. *Significant Effects:* Mild carbide-forming tendency. May contribute to certain porosity problems. Dry nitrogen gas not dissolved in iron.

PHOSPHORUS—*Principal Sources:* Certain pig irons, cast scrap, and steel. *Typical Amounts Observed:* Widely variable, .01-.80%. *Significant Effects:*

Source: Foundry Management and Technology, October 1977

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ments. Certain imported pig irons, however, may be produced from ores not so well selected and thus may introduce harmful effects on the structure and properties of cast iron.

Many of the investigations carried out overseas of the effects of certain elements in cast irons are made because of the presence of those elements in local iron ores and, hence, in pig irons and virgin steel. It therefore is imperative that suppliers of pig irons accurately and reliably identify their product sources so that foundrymen can be aware of possible changes in tramp element contents and can compensate for them.

The list of tramp elements originating in pig irons is extensive, but only a few tramp elements commonly are encountered. Phosphorus in pig iron, steel, or cast scrap was initially a contaminant in the iron ore that was not or could not be eliminated economically. As a result, a residual level of phosphorus is carried into the iron ore at a level dependent on the source of the ore and the methods used to produce the pig iron. In this country we are familiar with various grades of pig iron based on residual

Present as steadite, but also dissolved in ferrite to strengthen and harden the matrix. High levels increase fluidity, but increase brittleness.

SULFUR—*Principal Sources:* Coke, carbon raisers, cast scrap. *Typical Amounts Observed:* .02-.18% (or over). *Significant Effects:* Presence as sulfides-type depends on Mn-S ratio. Minimum amount appears necessary to control graphitization. (.04-.06%). Increased S reduces fluidity somewhat. Increases cell count and shrinkage. Over .15% has been reported to promote gas holes.

TELLURIUM—*Principal Sources:* Usually a ladle addition. Also may be present as alloy in steel or cast scrap. Used in core washes applied to localized areas. *Typical Amounts Observed:* .005%. *Significant Effects:* Tremendous increase in chill formation. Promotes Type D and mesh graphite. Effects reduced by inoculation and cerium treatments.

TIN—*Principal Sources:* Usually in ladle addition. *Typical Amounts Observed:* .10%. *Significant Effects:* About .06-.10%. used to promote pearlite matrix.

TITANIUM—*Principal Sources:* Present in pig irons. *Typical Amounts Observed:* .010-.06%. *Significant Effects:* Graphitizer up to .10%, increased chill over .25%, cell counts reduced, Type D graphite promoted. Renders iron more responsive to inoculation, however, and .03-.04% residual is desirable. Ties up nitrogen.

ZINC—*Principal Sources:* Entrapped as nonferrous material in cast scrap or steel. *Typical Amounts Observed:* ND[2]. *Significant Effects:* No effect on graphite size, shape, or distribution. Tends to promote ferrite in matrix.

[1]Effects and amounts referred to are influenced greatly by the presence of other elements. [2]ND: Not detected.

RESIDUAL ELEMENTS IN MALLEABLE IRON[1]

ALUMINUM—*Principal Sources:* Ferroalloys, nonferrous entrapped in cast scrap, residual element in steel. *Typical Amounts Observed:* .003-.02%. *Significant Effects:* Decreases nodule number by .005-.03% materials addition, increases nodule number over .05%. Mottling tendency increased over .05%. Increases rate of FSG and SSG.[2] Increases subcritical graphitization over .01%. Residual of .006% decreases fluid slag and slag reaction gas holes. Does not produce pinholes unless excess is used (.025%).

ANTIMONY—*Principal Sources:* Tramp elements in cast scrap or steel. *Typical Amounts Observed:* .005%. *Significant Effects:* Slight retarding effect on FSG and SSG. No significant effect on mottling. Decreases nodule count and raises hardness after annealing when present at .1% level.

ARSENIC—*Principal Sources:* Tramp element in cast scrap or steel. *Typical Amounts Observed:* .01%. *Significant Effects:* No effect of either increasing or decreasing mottling in white iron.

BISMUTH—*Principal Sources:* Usually a ladle addition, .002-.01%. *Typical Amounts Observed:* .001-.01%. *Significant Effects:* Mottling prevented by additions of .005 to .10% (depends on section size, silicon content, etc.). No effect on FSG or SSG if nodule count is maintained at 80-150/sq mm. "Spiking" prevented and feeding improved with .01% addition.

BORON—*Principal Sources:* Intentional addition, also present in certain pig irons, enameled scrap. *Typical Amounts Observed:* .0004-.004%. *Significant Effects:* Greatly increases nodule number to a maximum until stable carbides are formed (over .01%). Promotes FSG and SSG. Carbides are retained after annealing. Effect of .12% Cr offset by .003% B.

CALCIUM—*Principal Sources:* Present in ferroalloys. *Typical Amounts Observed:* ND.[3] *Significant Effects:* Not normally present in malleable iron. Promotes mottling.

CHROMIUM—*Principal Sources:* Alloy cast irons or steel; also as plated steel. *Typical Amounts Observed:* .02-.10%. *Significant Effects:* Strongly retards FSG and SSG. Lowers nodule number. Results in retained pearlite first, then massive carbides. Effect of CR reduced by high-speed annealing, which permits graphitization before extensive segregation of Cr to carbides can occur.

COPPER—*Principal Sources:* Nonferrous materials mixed with steel or cast scrap. *Typical Amounts Observed:* .10-.20%. *Significant Effects:* Increases rate of FSG and SSG, also increases nodule number somewhat. Reduces galvanizing embrittlement.

HYDROGEN—*Principal Sources:* Exposure to gases in melting and prior to solidification. *Typical Amounts Observed:* .0002-.0015%. *Significant Effects:* FSG retarded at .0005%, but effect depends on pretreatment and annealing atmosphere. Increases spiking damage from interdendritic porosity.

LEAD—*Principal Sources:* Nonferrous materials mixed with steel or cast scrap; also from fluorspar, possibly. *Typical Amounts Observed:* .002%. *Significant Effects:* Slight effect of retarding FSG and SSG and in reducing mottling.

MOLYBDENUM—*Principal Sources:* Steel scrap. *Typical Amounts Observed:* .03%. *Significant Effects:* Increases depth of chill and chill plus mottle. Prevents galvanic embrittlement. No effect on nodule number. Retards FSG and SSG.

NITROGEN—*Principal Sources:* Exposure to gases in melting, particularly arc melting. *Typical Amounts Observed:* .002-.015%. *Significant Effects:* Reduces nodule number. Retards FSG and SSG in range of .006 to .016%. Increases nodule size. Can cause worming pinholes. Side reactions with Al and Ti can neutralize effect of N.

OXYGEN—*Principal Sources:* Gases in melting, oxide slags. *Typical Amounts Observed:* .002-.005%. *Significant Effects:* Oxidizing melting conditions decrease nodule number and affect mottling, FSG, and SSG. Difficult to separate effect of oxygen from side reactions and oxide formation.

TELLURIUM—*Principal Sources:* Usually present as a ladle addition, but may be present in steel or cast scrap. *Typical Amounts Observed:* .004%. *Significant Effects:* Mottle prevented in heavy sections by .003 to .005% addition (depends on Si content). Decreases nodule number. No retardation of FSG and SSG at .01% Te if nodule count is over 80 to 150/sq mm.

TIN—*Principal Sources:* Tramp element in steel or cast scrap. *Typical Amounts Observed:* .007-.02%. *Significant Effects:* Both FSG and SSG are retarded over .02% Sn, but this can be counteracted by an increase in Si. Up to .08%, no effect on mottle or nodule count, shape, and size. Reduces nodule number beyond that level.

TITANIUM—*Principal Sources:* Present in pig irons. *Typical Amounts Observed:* .01-.02%. *Significant Effects:* Increases mottling and can raise nodule count. Contributes to graphitization because of side reactions with N, O,.

ZINC—*Principal Sources:* Entrapped as nonferrous material in cast scrap or steel. *Typical Amounts Observed:* ND.[3] *Significant Effects:* Anomalous and erratic effects, all of which are slight. It decreases nodule number at .02% and slightly increases FSG and SSG.

[1]Effects and amounts referred to are influenced greatly by presence of other elements. [2]FSG and SSG: First and Second-Stage Graphitization. [3]ND: Not detected.

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RESIDUAL ELEMENTS IN DUCTILE IRON[1]

ALUMINUM—*Principal Sources:* Residual aluminum in steel, aluminum entrained in steel or cast scrap, ferroalloys. More likely to cause pinholes if Ti is present. *Typical Amounts Observed:* .003-.06%. *Significant Effects:* Aluminum included in ferroalloys to suppress chill. Greater levels tend to cause short, stubby graphite flakes to form. Effect is greater in heavier sections. Literature notes that .005% may cause pinholes, but others report .01 or .10% used to eliminate pinholes. Effects may differ in base iron or ladle additions.

ANTIMONY—*Principal Sources:* Tramp element in cast scrap or steel. *Typical Amounts Observed:* .005%. *Significant Effects:* Very little effect on graphite spheroids, particularly when rare earths are present. Nodule degeneration at .25% Sb. Without rare earths, .02-.03% reported to cause severe loss of nodularity. About .05-.10% required to insure fully pearlitic matrix.

ARSENIC—*Principal Sources:* Tramp element in cast scrap or steel. *Typical Amounts Observed:* .01%. *Significant Effects:* No effect on nodule quality. About .10% required for fully pearlitic matrix.

BISMUTH—*Principal Sources:* Usually a ladle addition. *Typical Amounts Observed:* .01%. *Significant Effects:* With rare earths absent, Bi promotes flake-like graphite (.05% Bi), but improves nodule count and quality when sufficient rare earths are present.

BORON—*Principal Sources:* Certain pig irons, enameled scrap, intentional addition. *Typical Amounts Observed:* .0005%. *Significant Effects:* At about .005%, carbides form that are extremely resistant to annealing.

CALCIUM—*Principal Sources:* Ferroalloys. *Typical Amounts Observed:* ND[2]. *Significant Effects:* Optimum addition with postinoculant. Believed to increase nodule count and nodule quality. Excessive amounts promote primary carbides.

CHROMIUM—*Principal Sources:* Alloy steels and gray irons or as plated material in steel or cast scrap. *Typical Amounts Observed:* .02-.13%. *Significant Effects:* Very powerful carbide former. Carbides resistant to annealing. Must be below .04% for as-cast ferrite and generally no higher than .10% for pearlite.

COPPER—*Principal Sources:* Nonferrous materials mixed with steel or cast scrap. Also as ladle addition. *Typical Amounts Observed:* .01-.90%. *Significant Effects:* No effect on nodule quality or on primary carbides. Has significant effect on pearlite-forming tendency; about .50% yields fully pearlitic matrix, but depends on section size and other alloys present.

HYDROGEN—*Principal Sources:* Exposure to gases in melting and prior to solidification. Wet furnaces and ladle refractories. *Typical Amounts Observed:* .002-.0015%. *Significant Effects:* Said to promote inverse chill or centerline carbides. Also associated with pinhole formation.

LEAD—*Principal Sources:* Nonferrous materials mixed with steel or cast scrap. Also may be carried in with fluorspar, leaded steels, and copper additions. *Typical Amounts Observed:* .003%. *Significant Effects:* One of most potent subversive elements. Severe nodule degeneration caused. About .01% mischmetal has been said to counteract .05% Pb.

MANGANESE—*Principal Sources:* As an alloy in steel or cast scrap. *Typical Amounts Observed:* to 1.20%. *Significant Effects:* Intercellular carbides may be experienced over .70%. Generally less than .20% for ferritic grades. From .50 to .90% added for as-cast pearlitic grades.

MOLYBDENUM—*Principal Sources:* As an alloy in steel or cast scrap. *Typical Amounts Observed:* .0101-.50%. *Significant Effects:* Increases hardness by increasing pearlite forming effect. Mild carbide former, but segregates to cell boundaries if nodule count is low.

NITROGEN—*Principal Sources:* Exposure to gas in melting, particularly arc melting. *Typical Amounts Observed:* .002-.008%. *Significant Effects:* Mild carbide-forming tendency. May contribute to certain porosity problems. Dry nitrogen gas not dissolved into iron. Amount in base iron is reduced by Mg treatment.

OXYGEN—*Principal Sources:* Gases in melting, oxides in slags. *Typical Amounts Observed:* .005%. *Significant Effects:* Present, for the most part, as oxides that could be reduced by carbon to yield a gaseous product. Amount soluble, even in melt, is extremely small.

PHOSPHORUS—*Principal Sources:* Certain pig irons, cast scrap, and steel scrap. *Typical Amounts Observed:* .05%. *Significant Effects:* Forms phosphide network. Generally kept as low as possible (.06%). Raises ductile to brittle transformation temperature.

SULFUR—*Principal Sources:* Coke, carbon raisers, cast scrap. *Typical Amounts Observed:* .02%. *Significant Effects:* Reacts preferentially with magnesium or rare earths.

TELLURIUM—*Principal Sources:* Usually as a ladle addition. May be entrapped in steel or cast scrap. Can be present as alloy in some steel scrap. *Typical Amounts Observed:* .005%. *Significant Effects:* Considered a subversive element, causes some graphite degeneration when rare earths are not present. Has been added to control pinholes in ductile irons (1-2 g/100 lb.)

TIN—*Principal Sources:* Usually as a ladle addition. Also may enter through cast scrap. *Typical Amounts Observed:* .10%. *Significant Effects:* Powerful effect on pearlite formation. Over .10% Sn can result in nonspheroidal graphite in cellular boundaries.

TITANIUM—*Principal Sources:* Present in pig irons. *Typical Amounts Observed:* .07%. *Significant Effects:* Considered a subversive element, causes similar effect to that of aluminum, but stronger. Increased pinhole tendency if Al is present. Rare earths tend to overcome these effects.

VANADIUM—*Principal Sources:* Alloying element in steel or cast scrap. *Typical Amounts Observed:* .04%. *Significant Effects:* Forms very stable carbides.

ZIRCONIUM—*Principal Sources:* Present in some ferroalloys. *Typical Amounts Observed:* .01%. *Significant Effects:* Promotes flake graphite, causes spheroids to be less rounded.

[1]Effects and amounts referred to are influenced greatly by presence of other elements. In general, effects may be considered additive. For example, when copper is present, the deleterious effect of Sn occurs at a lower level of Sn.
[2]ND: Not detected.

phosphorus content.

Manganese similarly is present in most foundry-grade pig irons, carried over from the iron ore. The amount of manganese present in irons, however, usually is in excess of that assignable to pig irons.

Boron and titanium occur to various extents in iron ores. Depending upon the amount present in the ore and the techniques used to refine the ore to pig iron, the levels of boron and/or titanium in the iron will vary. In many instances, boron and/or titanium levels may be related to the silicon content of the pig iron for a given source and composition of the iron ore.

Cast Scrap—Foundrymen should keep in mind two factors involved in analyzing cast scrap as a source of tramp elements:

1. The type of cast scrap generally but not always dependably is a good indicator of the potential tramp elements that may be introduced. We have learned to recognize that cast iron radiators, for example, are commonly produced from high-phosphorus gray irons because of the improved fluidity imparted to the iron by the phosphorus. Engine blocks usually involve similar alloy additions. In recent years, however, some manufacturers have preferred to use tin additions to reduce the level of other pearlite stabilizers. As a result, categories of cast scrap may vary somewhat in the levels of tramp elements introduced.

2. A significant factor affecting the tramp element content of cast scrap is the technique used and dependability experienced in separating cast scrap from steel and nonferrous constituents associated with it.

In both of the above instances, foundrymen must depend upon their scrap processors and monitor their effectiveness with regular, routine inspection of incoming material.

Steel Scrap—The source of scrap and the proper separation and/or blending of scrap by the scrap processor determine the level of tramp elements introduced from this source. The bulk of all steel used in this country is still mild steel—selected for ease of fabrication, weldability, and adequate strength. Structural steels are changing in composition, however, as new high-strength, low-alloy steels are

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used in increasing quantities. Substantial quantities of these newer steels are capable of introducing certain tramp elements into cast irons.

It also is important to recognize alloy steels—steels designed with alloying elements selected to impart hardenability and strength. The scrap processor depends upon his ability to categorize scrap components of alloy steel to prevent these materials from entering general steel scrap. In the same manner, stainless and high-alloy steels are segregated and distributed to specialty markets. Again, monitoring of incoming scrap pays dividends to the foundryman.

Other Sources of Tramp Elements—Virtually everything added to cast irons during melting and processing represents a possible source of tramp elements. Ferroalloys are common sources of aluminum, calcium, zirconium, and titanium. While the levels of elements from these sources could reach values where they might be detrimental, the amount of these elements in ferroalloys is designed to be beneficial when appropriately used.

Ladle additions that are not properly controlled also may contribute to tramp element levels. The use of substitutes for pure ladle additions is a common source of difficulty. Impurities in coke and carbon raisers—particularly sulfur—are well known. Certain fluxes, when not properly used and when permitted to decompose thoroughly, also may represent a source.

Melting Methods Employed—The methods used for melting cast irons cannot be ignored in a discussion of tramp elements. Variations in the temperatures attained and time at critical temperatures directly affect the recovery of the more volatile tramp elements. With sufficient temperatures and adequate times, many of these tramp elements will be volatilized or "boiled off." Oxidation during melting also can affect the tramp element level, although the method of oxidation could introduce certain gaseous tramp elements into the system.

Typical Tramp Elements—The boxed data in this article show typical levels of tramp or residual elements in malleable, gray, and ductile irons. In some instances, the amounts observed are spread over rather broad limits that encompass effects ranging from negligible to severe. The boxes also indicate the most common sources of these elements and summarize the more significant effects they have on microstructure, properties, and casting characteristics.

The effects obtained are dependent on the presence of other elements in the system. The possibility of interaction between various elements must be recognized because those interactions could worsen or lessen the effect of a given element. For example, the use of cerium or rare earths to counteract the effect of certain tramp elements such as lead in ductile iron is well known, yet cerium in an iron produced from a relatively pure charge material can itself be detrimental, particularly in heavy-section castings.

In some instances anomalous effects are recorded in the literature concerning a given element. What one must keep in mind is that the effect of a given element depends on the amount of that element in the iron, the possibility of interaction with other elements, and the methods used to identify the effects observed. Unfortunately, the documentation of data regarding the effect of a given element is not always clear.

What To Watch For—The first thing to look into when a tramp element problem is suspected is to be certain that other procedural changes have not caused the observed effect. Carbide formation in gray irons, for example, may be induced by carbon equivalent control, metal temperature, inoculation effectiveness, and/or pouring rate—factors that often are more subject to change than the presence of tramp elements.

When a real tramp element problem occurs, it more commonly represents a sudden and drastic change from typical operating conditions, such as when a charge of lead-containing scrap is added to a furnace. In this case the effect occurs suddenly, may be assignable to only one furnace, and, depending on the type of melting system employed, may disappear as rapidly as it was noticed. At times, however, the effect of a tramp element may build up gradually and without notice, until suddenly a group of castings is found to be defective—for instance, when the titanium builds up in .malleable iron. It is much more difficult to identify the cause of problems in this case, and doing so may require considerable investigation.

Fortunately, if the foundryman is aware of the type of raw materials he purchases and is cognizant of the problems that can be experienced in the use of certain grades of scrap, the tramp element problem can be controlled adequately. Microstructural results and trends in mechanical properties often are as effective in identifying a tramp element problem as is chemical analysis, particularly since it is costly and time consuming to analyze for the host of elements that may prove to be detrimental.

The specific influence of a given element is not categorized easily since the effect and the effective amount of a given element may be influenced greatly by the presence or absence of other elements in the iron. Although some of these interactions have been identified, and guidelines set forth as a result of carefully monitored tests, many other possible interactions (positive and negative) are possible.

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Welding of Spheroidal-graphite Iron*

By D. R. Thorneycroft, B.Sc., A.I.M.

Whilst the main efforts of foundry technology are directed to the production of sound castings, welding processes are becoming reasonably regarded as legitimate methods for the reclamation of defective products. There is nothing deprecatory in this for foundry technology, as any basic production process will face a time-lag in "catching up," technically speaking, especially in conditions of rapid invasion of wider and more-stringent fields of application. Quality standards although necessary and inevitable, are easier to establish than the corresponding means of meeting them. The successful utilization of complementary techniques to achieve satisfactory castings is therefore sound practice both "ethically" and economically. Fabrication of castings by welding is also a growing practice that extends the purview of the foundry industry. The Author gives considered attention to the pearlitic, ferritic and austenitic grades of s.-g. iron and covers, in detail, the appropriate choice of welding technique, weld preparation, electrode and metallurgical factors that arise in their use. Bronze welding is considered and the special problems involved in the welding of acicular s.-g. iron are dealt with.

It is common practice to distinguish between the principal classes of spheroidal-graphite iron by reference to the metallurgical structure of the matrix. This classification may be summarized on this basis with the following five main types: Pearlitic, ferritic, acicular, martensitic, and austenitic s.-g. irons. There are, additionally, appropriate specifications which define the various grades of s.-g. cast iron, in terms of structure, mechanical properties and chemical composition. Two recent specifications in the United Kingdom are BS 2789: 1961 "Specifications for Iron Castings with Spheroidal or Nodular Graphite" and BS 3468:1962 "Specification for Austenitic Cast Iron," both being issued by the British Standards Institution. The bulk of the s.-g. iron produced is of the pearlitic, ferritic and austenitic grades, and the greatest welding interest is with these materials. Considerable attention has been given by metallurgists to their welding metallurgy, and it is now clear that the metallurgical factors involved are substantially the same as those encountered with similar flake-graphite irons. However, the pearlitic and ferritic s.-g. irons offer advantages because of their inherently-greater toughness and lower content of impurity elements, such as sulphur and phosphorus. Thus these grades are somewhat easier to weld by the metal-arc or oxy-acetylene welding processes, because less preheat is required. It must be emphasized that all except the austenitic grades of s.-g. iron are liable to excessive hardening in the heat-affected zones of welds, notably when arc-welding processes are employed, and for this reason, the welding response of the irons is not comparable to that of mild steel, the satisfactory metal-arc and gas-welding characteristics of which are a universal experience.

* Based on a lecture given to the national congress of the *Association Technique de Fonderie de Belgique* last, November, at Ghent University. The Author is welding development officer, International Nickel Company (Mond), Limited, Development and Research Department.

Application

Welding processes are becoming increasingly established in the foundry industry as legitimate methods of reclaiming defective castings, or for the repair of production defects, such as over-machining or even damaged castings. Provided faults are accessible, and not extensive compared with the size of the casting, it may become economical to reclaim a casting by welding. As s.-g. iron castings are characterized by high strength and toughness, and certain grades may also have other special properties[1,2] it becomes necessary, therefore, to consider whether the effects of welding will impair service performance of these high-grade materials.

Apart from foundry welding to reclaim defective or damaged castings, the various s.-g. irons may require to be welded to themselves, or to steels, in the production of components or in the fabrication of structures. For instance, ferritic s.-g. iron castings are welded to mild steel and stainless steels, and austenitic s.-g. iron castings to stainless steels, employing either the metal-arc or argon-shielded tungsten-arc processes. However, when welding leads to the formation of hard zones in the casting, it may be necessary to restrict fabrication to those applications which do not involve severe impact loading, or high fluctuating stress.

Choice of Process

The choice of welding processes for s.-g. irons is usually between metal-arc or oxy-acetylene welding or, in some circumstances, is dictated by the availability of equipment. Metal-arc welding is the process which to date has been most widely applied to the welding of the s.-g. irons, since, with this process, distortion due to the heat of welding is

Source: Foundry Trade Journal, April 25, 1963

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relatively slight and pre-heating is at relatively-low temperatures. However, oxy-acetylene welding is also used and in special cases argon-shielded welding and even resistance welding have been successfully employed. The application of resistance welding by the flash process has been mainly to pearlitic or ferritic s.-g. iron pipe for the purpose of making pressure-tight butt-joints, and there is scope for extending the process to other components. Although the technique is largely in the development stage, the indications are that hardening and undesirable structural change can be less marked than in metal-arc welds[3]. Most of the molten cast iron is expelled from the weld interface during the upset stage of the operation, and in a correctly-made weld very little of embrittling eutectic-carbide will be present. The Swedish Welding Institute have described the application of the process to cast irons in general[4], and comment favourably on the technique for such materials presenting inherent welding difficulties. There is little application of bronze-welding to the pearlitic or ferritic s.-g. irons but some aspects of the process are described later.

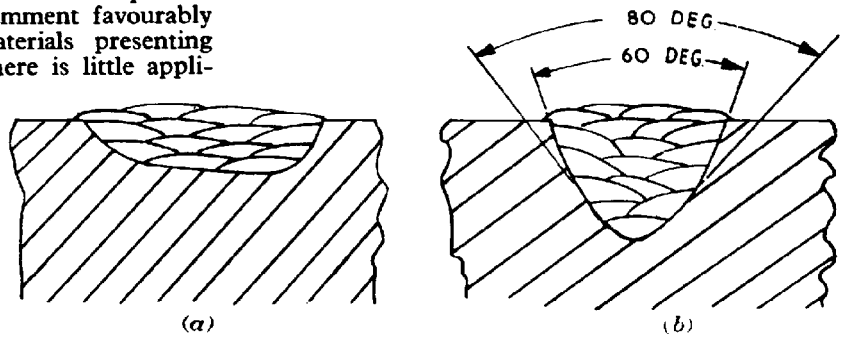


FIG. 1.—*Some examples of preparation and welding procedure on castings with various typical defects.*

Key: (*a*) Preparation and procedure for shallow defect, (*b*) for deep-seated defect and (*c*) for corner defect. Casting preferably to be positioned for downhead welding.

Preparation

Irrespective of which welding process is selected it is vitally important to prepare the casting adequately for the welding operation. This aspect has been carefully explained in the Codes of Practice for flake-graphite irons, and the recommendations made there[5, 6] apply equally to s.-g. iron castings. Thus removal of such foreign materials as oils, greases, paint marks, rust and sand from the surface in the weld area is essential since these contaminants will impair the quality of the weld; the casting skin must also be removed by grinding or machining. These precautions are equally necessary in preparation for reclamation welding or in producing welded components. Where rectification of a mis-run is involved it is again essential before depositing weld metal to remove the casting skin by grinding. Defective material appearing at the surface must be removed by grinding to obtain a sound base for welding and, where there is any doubt, the extent of the defective area is best found by radiographic examination. Shallow or deep grooves may be formed as a result of these grinding operations, as sketched in Fig. 1. Where defects are located in relatively-inaccessible positions, it is necessary to remove sufficient material to permit a satisfactory welding operation. Where sub-surface defects are revealed only during final machining operations on the casting, it is necessary to remove all unsound material before attempting a repair.

Fabrication

Structural and component welding often requires precise joint preparation and therefore involves machining operations. The joint should be designed so that there is sufficient space to manipulate the electrode or filler rod and at the same time obtain proper fusion of the weld metal with the casting. Butt-joints for metal-arc welds are wider than those for mild steel (but similar to those for flake-graphite irons); the included angle being 60 to 75 deg. for V-joints and 20 to 30 deg. for U-joints—a root-radius of five to seven mm. is preferable. To minimize distortion, a double V-joint is recommended whenever possible in welding heavy sections. Single V- or double V-joints with an included angle of 90 deg.—the more usual preparation for full-penetration gas-welds—follow the normal practice for grey cast irons. In subsequent sections, the welding characteristics of each class of s.-g. iron is presented in detail.

PEARLITIC AND FERRITIC S.-G. IRONS

The pearlitic and ferritic irons make up the largest group of s.-g. irons produced, and, being well-established engineering materials, their welding response has received most attention. Whilst little or no practical guidance on welding procedure has previously been published, several accounts are available of the metallurgical characteristics of welds, particularly of the structural changes produced by the heat of a welding operation. An excellent survey of the subject is given by G. R. Pease in an inter-

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TABLE 1.—*Details of Weld Metal deposited from 55-per-cent. Nickel/Iron Electrode.*

Element ...	C	Si	Mn	Mg	Ti	S	Fe	Mi
Percentage ...	1·3	0·3	0·32	0·01	0·22	0·006	43·8	54·0 (by difference)

Mechanical Properties

Condition	Yield strength, (tons per sq. in.).	Maximum strength, (tons per sq. in.).	Elongation on 1½ in. (per cent.).	Reduction of area (per cent.).	Hardness (DPN).	Unnotched Charpy impact value (kg.-m.).
As-deposited	28 27	37½ 36	9 7	9 13	180 —	11 10
Heat-treated 2 hr. at 850 deg. C.	30	35	7	14	—	—
Plus 8 hr. at 700 deg. C. ...	27½	35	6	9	200	15

pretative report of the American Welding Research Council[7]. It is generally observed that the most marked changes in structure are obtained in the heat-affected zone of electric-arc welding, but the metal-arc process is often preferred since distortion due to the heat of welding is relatively slight and preheating is at relatively-low temperatures.

Because of the unique mechanical properties of the above s.-g. irons, consideration has had to be given to the use of electrodes or filler rods in the production of welds of similar strength and toughness; this requirement has led to the development of new welding materials. Many foundrymen will be familiar with the flux-coated non-ferrous electrodes* which are so widely employed for the metal-arc welding of flake-graphite grey cast irons. Whilst these electrodes can be used to make sound joints in s.-g. irons the resulting joint strength is not always adequate and application of a 55-per cent. nickel/iron electrode is usually recommended, the strength of the weld metal being comparable with that of the iron.

Electrodes

Details of weld metal deposited from the 55 per cent.-nickel/iron electrode are given in Table 1. Most electrode manufacturers produce a flux-coated nickel/iron electrode, but it is advisable to check whether they are suitable for ac or dc operation, and can be used with the welding equipment available. A further requirement is that the graphite of the weld deposit shall be mainly spheroidal in form. To ensure maximum freedom from porosity in weld deposits most nickel/iron electrodes should be re-baked for at least one hour at 250 deg. C. in a well-ventilated electric oven and either used immediately or stored in a similar oven at 120 deg. C. until required. The choice of 10-, 8- and 6-gauge electrodes will usually satisfy most welding requirements. The welding current should be kept as low as possible consistent with obtaining smooth operation and a good wash at the sides of a joint. To ensure the latter feature the use of a slight weaving technique is recommended, although, when weaving, the width of the deposit should not exceed three times the diameter (*i.e.*, core-wire diameter) of the electrode. Whenever possible the casting should be positioned for downhand welding since this procedure facilitates all aspects of the welding operation. Where extra-long welds, or several repair positions are involved, it is preferable to stagger the welding operation to distribute the heat and to minimize distortion.

FIG. 2.—*Microstructure at the weld junction of a metal-arc weld in ferritic s.-g. iron—as-welded.* × 300.



* Under the proprietary name of Monel, nickel/copper, or plain nickel.

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Welding of Spheroidal-graphite Iron

Slag

The nickel/iron electrode usually gives no problems with slag removal, which can be easily accomplished with a steel chisel or even by wire brushing if the weld beads have been laid with a satisfactory profile. It is essential to remove the slag from each crater before making a re-strike and to remove it completely from each weld-run before depositing the adjacent weld. Further to ensure maximum weld soundness, the forward movement of the electrode tip should be accelerated as the end of the weld-run is approached; this will taper the run rather than ending it abruptly in a large weld pool.

Pre-heating

Experience with the welding of the pearlitic grade of s.-g. iron[7,8] has indicated that it is desirable to pre-heat the casting to 200 to 250 deg. C. (particularly if high welding stresses are anticipated) to avoid cracking in the heat-affected zone. Pre-heat is normally not required for the wholly-ferritic grade of iron. The method of preheating will be governed by such factors as size of casting, and availability of furnaces or of gas-heating appliances. Since s.-g. iron is strong and relatively ductile, it is often sufficient to preheat locally rather than to heat the complete casting, as would normally be required with most engineering grades of flake-graphite cast iron. Small castings or thin sections may need to be heated continually since heat is quickly lost to the surroundings, even before a welding operation has been completed. Because of these moderate preheating requirements, the pearlitic s.-g. irons are somewhat easier to weld than the corresponding flake-graphite irons.

Cooling

To guard further against cracking in the heat-affected zone of the pearlitic grade of material it is generally advisable to cool as slowly as possible after welding although, in most cases, it is sufficient merely to cool under a cover of heat-insulating material such as asbestos, sand or ashes. Alternatively, the welded casting may be charged direct to the furnace if a subsequent annealing treatment is to be given.

Provided the temperature of the surroundings is not below 20 deg. C. the ferritic grade of s.-g. iron is weldable by the metal-arc process without preheat. However, there would be some risk of cracking when site-welding under wintry conditions, or in unheated shops during such periods, and in such circumstances a preheat is advised. Localized preheating does ensure that surfaces adjacent to the weld, or to be incorporated in the weld, are free from moisture or from volatile contaminants which will otherwise enter the weld pool and cause porosity. This effect is sometimes the cause of lack of pressure-tightness on testing of castings incorporated by welding into, for instance, hydraulic components in certain assemblies for welding.

Structural Changes

The heat of the metal-arc welding operation will induce a structural change in the heat-affected zones of both ferritic and pearlitic grades (in accordance with the well-known transformation characteristics of ferrous materials), and in this respect the s.-g. irons are not different from the flake-graphite irons. Thus even with modest preheat, rates of cooling in

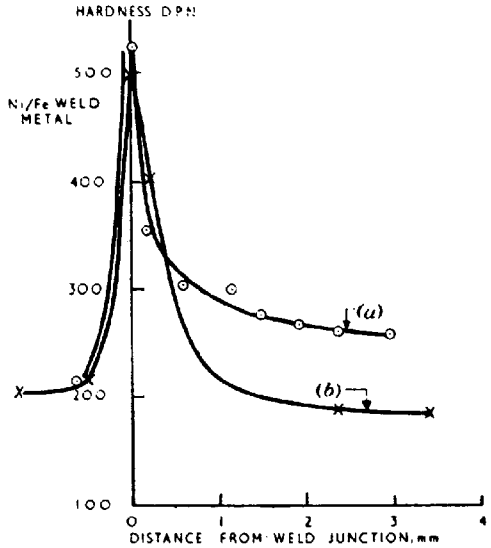


FIG. 3.—*Hardness surveys across metal-arc welds in s.-g. iron with no post-weld heat-treatment.*

Key: (*a*) S.-g. pearlitic iron and (*b*) s.-g. ferritic iron.

the heat-affected zones of a metal-arc weld are sufficiently rapid to cause the formation of hard transformation products, notably of eutectic-carbide and martensite (Fig. 2). These hard products together form a narrow, brittle zone at the junction between weld-metal and casting and its presence is revealed, for instance, by the peak value on a hardness survey across the joint; Fig. 3 illustrates two such hardness surveys, which are typical for as-welded joints in the pearlitic and ferritic grades of s.-g. iron. Nevertheless, such welds are reasonably strong, and moderately tough and ductile. Examples of the tensile properties of butt-welds made in 1-in. thick castings with a 55-per-cent.-nickel/iron electrode are given in Table 2. The toughness of ferritic s.-g. iron is markedly affected in the weld zone by metal-arc welding, although post-weld heat-treatment is beneficial.

Mild-steel/S.-g. Iron

Welded joints involving mild steels and either the pearlitic or ferritic grade of s.-g. iron are being encountered as the potentialities of these high-strength irons are recognized in welded structures. Metal-arc welding with a nickel or nickel/iron electrode is the least difficult method of making the joint between steel and s.-g. iron and the process offers particular advantages over gas welding when the mild-steel member of the assembly is used in the cold-worked condition, for the purpose of high strength, since annealing can be expected in the vicinity of a gas-weld. The tensile strength of the joint is usually greater than that of annealed mild

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steel, but some embrittlement is to be expected in the heat-affected zone of the casting. The toughness of the joint is sufficient in applications that do not involve frequent stressing or high-impact loads, one such construction is shown in Fig. 4. Mild-steel arc-welding electrodes are not suitable for use with the steel s.-g. iron combination because risk of cracking in that portion of the weld adjacent to the s.-g. iron is high; similarly the use of aluminium-bronze welding electrodes is not recommended .

Oxy-acetylene Welding

This process finds application in welding small and large castings when the weld zone is required to have mechanical properties and corrosion characteristics matching as closely as possible those of the component; it is a type of welding that has to date been employed mainly in the foundry in the reclamation of defective castings, using s.-g.-iron filler rods which for the most part have been experimental compositions. In general, individual foundries have tackled the problem of developing a suitable filler rod of the normal s.-g.-iron composition, using cerium-* or magnesium-treated iron. However, welds made with a filler rod of normal s.-g.-iron composition tend to be excessively brittle in small sections, or porous and low in strength in large sections.

A filler rod has been developed[9] containing the following (percentages), 3.75 C, 2.75 Si, 5 Ni, <0.06 Al and 0.06 Mg (maximum), which largely overcomes the problems of porosity and low strength. This rod gives rise to sound, strong, welds in the pearlitic grade of s.-g. iron which, in the as-welded condition, have tensile properties comparable with those of the casting; typical data

* Available from Sandholme Iron Company, Limited.

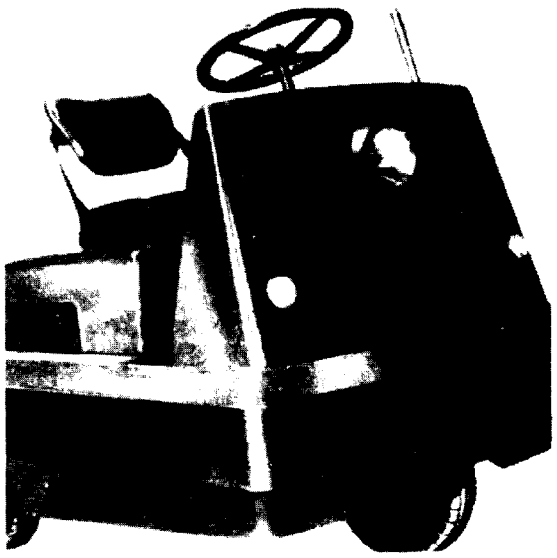


FIG. 4.—*Battery electric tractor with ferritic s.-g. iron frontal casting incorporated in the mild steel chassis by metal-arc welding. (Lansing Bagnall, Limited.)*

TABLE 2.—*Tensile Properties of Butt-welds made in One-inch-thick Castings with a 55-per-cent. Nickel/Iron Electrode.*

Initial structure.	Condition of the casting.	Tensile properties. Yield strength, tons per sq. in.	Mean strength, tons per sq. in.	Elongation on 2-in. per cent.	Reduction of area per cent.	Location of fracture.
Pearlitic ..	Unwelded	38	40	3	nil	—
	As-welded	34	37	2	4	Weld junction
		32	37	1	3	Weld junction
	Welded, plus 3 hr.,	22½	27	7	9	Weld metal
	900 deg. C.	22½	29	12	15	Casting
	plus 16 hr., 700 deg. C.	22	25	6	7	Weld metal/weld junction
Ferritic ...	Unwelded ..	21	30½	18	23	—
	As-welded	25	28½	4	16	Casting
		24	28½	7	11	Weld metal
	Welded, plus 2 hr., 850 deg. C.	22½	30	6	3	Weld junction/weld metal
	18 hr., 700 deg. C.	17½	30	7	5	Weld junction/weld metal

Unnotched Charpy impact-strengths of above butt-joints:

	Unwelded.	As-welded.	Heat-treated.
Pearlitic casting	10 ft.-lb.	13 ft.-lb.	38 ft.-lb.
Ferritic castings	100 ft.-lb.	16 ft.-lb.	35 ft.-lb.

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for ½-, 1- and 2-in.-thick castings are set out in Table 3. Furthermore, tests on weld metal deposited by this filler rod reveal the high strength (in combination with reasonable ductility) that can be obtained in the annealed conditions (see Table 4). The filler rod would therefore appear suitable for making welds in castings which are intended for use in the annealed condition; in such circumstances, welding should precede heat-treatment.

Gas-weld Procedure

The procedures to be followed in preparing castings for welding have already been described. It is also important when gas welding to remove the defective zone of the casting by grinding to provide sound metal on which to deposit weld metal; in many cases this procedure improves access to the area for the welding operation. In controlling the quality of welds the surface cleanliness of the filler rod is another important factor. Porous welds will result from the use of rods which are rusty or are imperfect due to an excessive casting skin; it is preferable to grind such rods lightly before use[10]. Use of a proprietary brand of gas-welding flux (of the kind normally used for the flake-graphite grey cast irons) is acceptable, but the flux should be applied sparingly in conjunction with a soft, neutral welding flame. As a rough guide jet size should be similar to that normally selected for a similar steel section, but the size of flame should be chosen, after a trial, to provide just sufficient heat to maintain a molten weld pool in which the filler rod can be melted by immersion and not by direct contact with the flame. To ensure maximum freedom from porosity, turbulence, like that caused by vigorous puddling or by harsh flames, must be avoided.

The heat-affected zone of an oxy-acetylene weld is usually wider than that obtained with metal-arc welding. The heating and cooling during and after the welding operation has the effect of producing, in the pearlitic grade of iron, a heat-affected zone with a structure similar to that of the original casting. On the other hand, if a ferritic casting is found to require welding, either at the machining stage subsequent to heat-treatment, or possibly in component welding, then both weld deposit and heat-affected zone will have substantially pearlitic structures; the full ductility characteristic of the ferritic grade can be restored only by a further annealing treatment. Where a pearlitic casting is to be annealed for service it is desirable to carry out welding operations on the casting before application of the annealing treatment.

BRONZE WELDING

The bronze-welding process is a high-temperature brazing operation which involves no fusion of the base iron and the risk of a marked structural change occurring in the heat-affected zone is consequently less. It is well known as a satisfactory procedure for the repair-welding of the relatively-brittle flake-graphite grey cast irons, but, when applied to the joining of s.-g. irons some difficulty is encountered

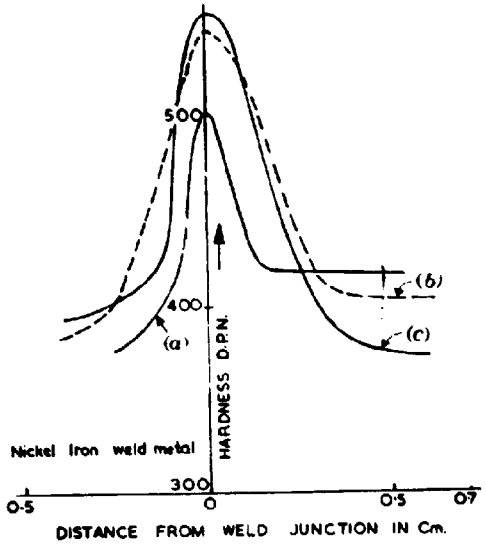


FIG. 5.—*Hardness surveys across metal-arc welds in acicular s.-g. iron.*

Key: (*a*) No pre-heat but finally tempered at 350 deg. C. for three hours; (*b*) no pre-heat and (*c*) pre-heat 320 deg. C.

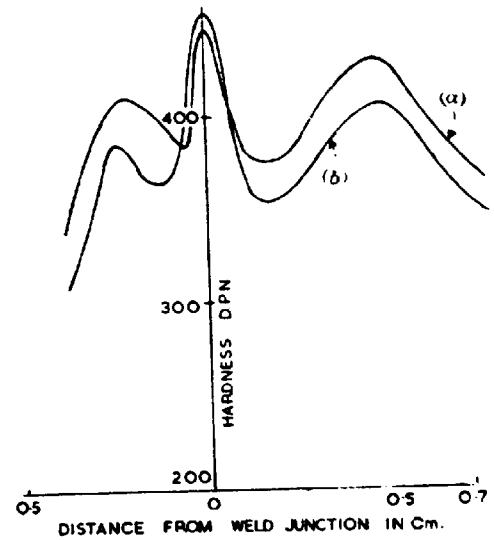


FIG. 6.—*Hardness surveys across oxy-acetylene welds (using five-per-cent.-nickel/s.-g. weld metal) in acicular s.-g. iron.*

Key: (*a*) Tempered at 350 deg. C. for three hours and (*b*) as welded.

in obtaining consistently, high bond strength at the joint interface (a requirement which is, of course, essential because of the high strength of the casting). Lack of consistent bonding appears to be a problem of wettability, a problem which is partly resolved by the use of special welding fluxes, for instance, in England, that produced by the Suffolk Iron Foundry (1920), Limited, is sufficiently aggressive to aid wetting. A definite advantage in joint strength is to be gained by the use of nickel bronze welding rods; one specification is (percentages) 41

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to 45 Cu, 14 to 16 Ni, 0.2 to 0.5 Si, 1.0 Sn (maximum), balance Zn.

Preparation

Experience indicates that graphite must be removed from the face exposed for brazing, and treatment of the machined or ground surfaces by vapour-blasting procedures using fine grit (70 mesh) or by swabbing with a cold 10 per cent. solution of sulphuric acid appear to be effective. Whilst an aggressive gas-welding flux alone will remove graphite spheroids from the interface of the joint, these are then entrapped in the weld pool and gaseous reaction products involving carbon are formed with the result that the welds are porous and consequently lack strength. The maximum strength obtainable with reasonably sound and well-bonded joints is approximately 30 tons per sq. in. in pearlitic or ferritic s.-g. iron.

ACICULAR S.-G. IRON

Acicular spheroidal-graphite iron constitutes a relatively small tonnage of the total s.-g.-iron production; the composition of these special irons is carefully balanced (by the presence of 1.5 to 2 per cent. nickel and 0.7 to 1 per cent. molybdenum) to obtain transformation characteristics in the matrix which lead to the development of relatively hard, but desirable, structures (bainite) during cooling in the mould. The hardness of such material is usually in the range 280 to 350 BHN, but in castings welded by the metal-arc or the oxy-acetylene processes, zones of even higher hardness are formed at the edge of the weld and in the heat-affected zone of the casting.

Variations in hardness encountered in experimental welds produced using the 55-per-cent.-nickel/iron electrode for metal-arc welding, and the five-per-cent. nickel s.-g. iron filler rod for oxy-acetylene welding are shown in Figs. 5 and 6. It can be seen that the thermal cycles accompanying welding markedly affect the original condition of the casting. The most prominent feature, the presence of a narrow band of eutectiferous carbide at the weld interface in both the metal-arc and oxy-acetylene welds, causes the peak hardening observed in the hardness surveys. No reasonable conditions of preheat are found to reduce hardening in the weld zones, but the metal-arc welds are somewhat softer after a tempering treatment at 350 deg. C., although this treatment does not decompose the narrow carbide band at the weld interface. The tests have indicated that cracks are prone to develop in the heat-affected zone of the casting during metal-arc welding and, on the whole, the oxy-acetylene welding process is the more suitable technique.

AUSTENITIC CAST IRONS

The heat- and corrosion-resistant cast irons are finding increasingly wide application and welding procedures are required both for reclamation work in the foundry, and for fabrication purposes. Reference will be made primarily to the s.-g. Ni-Resist series. Full details of this series will be found in a recent survey[11] from which the compositional limits are given in Table 5. The Ni-Resist

TABLE 3.—*Tensile-test Results for Butt-welds in Pearlitic S.-g. Iron., made by Oxy-acetylene Welding with five per cent. Nickel Filler Rod.*

Thickness of casting (in.).	Type of weld	Size of tensile test-piece (in.).	Field Strength (tons per sq. in.).	Maximum Strength (tons per sq. in.).	Elongation on 50mm. (per cent.).	Location of Fracture.
½	Single-V (included angle 90deg.)	1 × 0·45	33½ 40 38	46½ 46½ 42½	2½ 2½ 2	centre of weld centre of weld weld junction
1...	Double-V (included angle 90deg.)	1 × 0·875	36 31 37	37½ 37 40	2 2½ 2½	weld junction weld junction weld junction
2...	as above ...	1 × 1·1 (axis of joint)	—	38	3	centre of weld
		0·505 dia. (surface of joint)	36 35½	42 41	1½ 1½	centre of weld centre of weld

TABLE 4.—*Mechanical Properties of Weld Metal as in Table 3 but relating to the Annealed Condition.*

Five per cent. nickel/s.-g.-iron weld metal annealed three hr. at 900deg. C. plus 16 hr. at 700deg. C.	Elastic limit tons per sq. in.	Maximum strength. tons per sq. in.	Elongation on one inch (per cent.).	Hardness DPN.
	36½ 36	39 41	8½ 12	230 230

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Welding of Spheroidal-graphite Iron

cast irons embrace a wide range of nickel contents, and with additions of chromium, manganese and silicon, present a variety of compositions. A comprehensive study of the welding response of these irons has probably never been attempted, and published information on welding characteristics is confined to experiences with Ni-Resist Type D2 and Type D2C. Whilst the concern of this paper is with s.-g. irons, the form of the graphite in the Ni-Resist series, *i.e.*, whether flake or spheroidal, does not appear to have a direct influence on welding response. However, sulphur and phosphorus are elements known to impair welding response, and maximum contents of 0.015 per cent. S and 0.05 per cent. P are recommended to reduce welding difficulties. An excess of magnesium may impair welding response, and it would seem desirable to restrict the residual content to the range 0.05 to 0.08 per cent. to achieve optimum graphite form and weldability. Manganese is apparently beneficial, approximately two per cent. is desirable; some of the Ni-Resist irons have a relatively-high manganese content.

Welding Practice

Limited information on welding practice reveals the following situation. Special welding electrodes are not commercially available for the Ni-Resist series of austenitic cast irons and metal-arc welding with the 55 per cent. nickel-iron electrode (*i.e.* the type of electrode normally employed when welding pearlitic and ferritic s.-g. irons) is the procedure most widely used; but it is necessary to preheat the casting (see below). The weld metal from this electrode is tough at room-temperature, would be expected to be tough at sub-zero temperatures, but is magnetic. The weld metal is not heat-resistant, and therefore does not compare with the heat-resistant, chromium-bearing Ni-Resist irons. More heat-resistant weld metals are those deposited from 25Cr/20Ni austenitic stainless-steel electrodes, and from high-nickel-chromium/iron welding electrodes such as Inco-Weld "A" or Inconel 182. The non-magnetic grades of Ni-Resist (type D2 and D2C) have been welded with the austenitic electrode to obtain non-magnetic weld zones.

Argon-arc

Argon-shielded tungsten-arc welding has been successfully described in an application in the aircraft industry[12], and this process is likely to be most suitable for the joining of thin-walled castings, or of such castings to sheet of wrought alloys. A nickel-iron filler wire (for instance, Nilo 55 produced by Henry Wiggin & Company, Limited, Hereford, England) produced satisfactory welds between type D2 Ni-Resist and a variety of wrought alloys, including 13 per cent. Cr steel, 25Cr/20Ni stainless steel and Nimonic 75. Filler rods of composition matching that of the casting are unsatisfactory because of weld metal cracking, although preheat at 600 deg. C. will eliminate this feature.

Oxy-acetylene

Oxy-acetylene welding is not widely used, partly because of distortion in thick sections, but additionally, cracking at the weld interface is a problem, which is currently being investigated on Ni-Resist type D2 and type D2C castings, using filler rods of modified composition. The results obtained to date suggest that the process may be suitable for the reclamation of castings. Porosity in weld deposits was substantially eliminated by the use of a reducing flame, no marked advantage was obtained from the use of fluxes.

Cracking

The major problem encountered when welding Ni-Resist cast irons has been cracking at the weld junction and in the heat-affected zone adjacent to the weld. Such cracking apparently occurs at high temperatures during the thermal cycle of the welding operation. Apart from the suggestion made earlier, to control the level of the impurity elements, sulphur and phosphorus, some benefit may be obtained from preheat, when metal-arc welding. Preheating to at least 350 deg. C. is recommended for castings of thick-section. The high-expansion Ni-Resist cast irons (for instance, type D2 and D2C) would be expected to show more distortion during heating and welding than the low-expansion Ni-Resist irons, containing 36 per cent. nickel, but there is insufficient practical experience to indicate the relative welding response of the various grades. Preheat will tend to reduce thermal

TABLE 5.—*Compositional Limits for Various Grades of S.-g. Ni-Resist Cast Iron.*

* Approximate equivalent grades in BS 3468 are, for 1 and 2—AUS 202; for 3—AUS 203, and 4—AUS 205.

Grade of S.-g. Ni-Resist.	Normal Composition, per cent.							
	C Maximum.	Si	Mn	S	P	Ni	Cr	Mg
*1. Type D2 ...	3·0	1·7–3·0	0·7–1·0	0·01	0·05	18–22	1·7–2·5	0·08
*2. Type D2B ...	3·0	1·7–3·0	0·7–1·0	0·01	0·05	18–22	2·7–4·0	0·08
*3. Type D2C ...	2·9	2·0–3·0	1·8–2·4	0·01	0·05	21–24	0·5 max.	0·08
*4. Type D3 ...	2·6	1·5–2·8	0·5 max.	0·01	0·05	28–32	2·5–3·5	0·08
Type D3A	2·6	1·5–2·0	0·5 max.	0·01	0·05	28–32	1·0–1·5	0·08
Type D4	2·6	5·0–6·0	0·5 max.	0·01	0·05	29–32	4·5–5·5	0·08
Type D5	2·4	1·5–2·75	0·5 max.	0·01	0·05	34–36	0·1 max.	0·08
Type D5B ...	2·4	1·3–2·75	0·5 max.	0·01	0·05	34–36	2·0–3·0	0·08

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gradients, and wherever possible, residual welding stresses should be removed by applying a post-weld stress-relief heat-treatment; usually by heating in the range 620 to 680 deg. C.

Acknowledgment

The Author is indebted to the International Nickel Company (Mond), Limited, for permission to publish this paper.

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Maintenance and Repair Welding

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1. HARDFACING

Introduction

In a previous issue of *Svetsaren* No. 3, 1977, pp 8–11, a paper was published which attempted to describe the scope of Arc Welding Surfacing Techniques in the maintenance and repair of worn components. At the same time the writer made a plea for a new design concept in which components subjected to wear should be designed in such a way that the worn parts could be resurfaced economically after a predetermined amount of wear had taken place. Thus the initial manufacturing process could make use of relatively cheap base material with the areas exposed to wear being clad with a more expensive wear resistant surface by one of the arc welding processes, the materials being chosen to give optimum wear resistance compatible with reasonable weldability and the possibility of renewing the welded surfaces several times. Such a procedure would save time, money and strategic raw materials such as chromium, molybdenum, nickel, cobalt and tungsten. In order to be successful the correct materials must be selected and since for a particular wear situation there will be one material or group of materials that will give optimum performance, it is important to identify the type of wear as accurately as possible.

Wear Mechanisms

The types of wear encountered in industry can be divided up as follows:[1]

Abrasive	50%
Adhesive	15%
Erosive	8%
Fretting	8%
Corrosive	5%

Often two or more types of wear are involved simultaneously or one type may change into another as wear proceeds.

Abrasive Wear (Grinding and Gouging wear)

Abrasive wear is caused by hard protrusions or particles penetrating and deforming a surface, eventually displacing slivers or chips from it to produce wear debris. This type of wear occurs when rock, sand or other hard substances slide under pressure across a surface. Thus ploughs, cultivators and bulldozer blades which make contact with soil, sand and stones are subject to abrasive wear. Machines for crushing ore and grinding minerals are also subject to abrasive wear. During abrasion a metal becomes work hardened and the hardness of the abrasive particles must exceed that of the work hardened metal surface if cutting is to occur. If the surface work hardens to a value equal to or exceeding that of the abrasive particles the latter become blunted and tend to deform the surface rather than cut it. In general hard surfaces resist abrasion better than soft ones, but if the abrasive particles are not too coarse and the surface is fairly soft, it can happen that the abrasive particles become embedded in the surface thus increasing its wear resistance particularly in the case of sliding abrasion. Increasing surface hardness in such circumstances can make things worse since it may prevent the abrasive particles becoming embedded in the surface.

Adhesive Wear (Metal to Metal wear)

Adhesive wear is a wear process which arises from the relative movement between two conforming metal surfaces. Seizure, galling, scuffing etc. are terms commonly used to describe adhesive wear. All surfaces, even polished ones, have projections or asperities when examined in cross section at high magnification. When the surfaces are in contact and relative movement occurs very high local pressures occur causing the tips of the asperities to deform and bonding to take place. These bonds are broken by the sliding forces so material is torn from one surface and transferred to another. In extreme cases welding can occur which prevents further movement (seizure). Adhesive wear is very much a function of the environment, operating conditions, temperature and pressure etc.

Erosive Wear

Erosive wear is a form of abrasive wear caused by fluid containing solid particles flowing over a surface. Relative motion between the surface and the fluid causes loss of surface material.

If the fluid flows along the surface so that motion of the solid particles is parallel to it, the term abrasive erosion is used. If the motion of the particles is normal to the surface, impingement erosion is obtained.

Fretting

Fretting is the name given to the wear produced when two surfaces in contact have an oscillatory motion relative to each other, the amplitude being small. The motion disrupts any protective oxide film formed on the surface and the resulting debris can act as an abrasive and make the situation worse. This is typical of the type of wear which can best be combated by changing the environment such as eliminating the source of vibrations, ensuring adequate lubrication etc.

Corrosive Wear

Corrosive wear (also called chemical wear) occurs in situations where sliding or abrasive wear takes place in an environment which reacts chemically with the wearing surfaces. The presence of an aggressive environment can greatly accelerate wear rate and it is often cheaper to try to improve the environment rather than change the nature of the surfaces.

This brief introduction to the different types of wear gives a very simplified picture of a very complex subject. A particular wear situation may not fit comfortably into any of the classes just presented or it may be a combination of two or more. In some situations it will be cheaper and more effective to change the environment or design rather than the material. For

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example if the blades of an air extraction fan wear out too quickly and investigation shows that this is due to the presence of sharp gritty particles in the air, it will probably be better to place a filter over the air intake to the fan rather than try to find a harder, more wear-resistant material for the blades.

In other situations the best solution may be to apply a surface coating to the areas where wear is occurring. Although there are many types of surface layer and many ways of applying them, this paper is only concerned with the relatively thick layers applied by an arc welding process.

The factors governing the choice of welding process have been dealthwith in a previous article[2], so the rest of this paper will concentrate on the selection of suitable filler material compositions for a variety of common wear situations.

Electrodes for Manual Metal-Arc Welding

The coated manual metal-arc welding electrode offers the widest choice of weld metal compositions because of the possibility of alloying via the coating. For certain materials, such as grey cast-iron, manual metal-arc welding is in practice the only solution because of the low heat input.

In repair work it is often necessary to weld 'in-position', particularly in the case of large objects, such as excavator buckets or other heavy components which cannot be positioned for welding in a flat position. Again, manual metal-arc welding offers the best solution.

Obviously, if the workpiece is of a suitable shape and there are many similar pieces to be welded, e.g. tractor tracks, some form of mechanized welding should be used.

Before beginning a repair weld all defective material must be removed. No attempt should be made to weld over cracks since these will only propagate when the part is put back in service. A convenient way of removing cracks and preparing for the repair weld is to use a cutting and gouging electrode such as OK 21.03, *Fig. 1.*

Repair welding often has to be carried out on material which was not originally selected for its good weldability. Again, manual metal-arc welding is often the only answer to such a problem, for the hand welding welder has

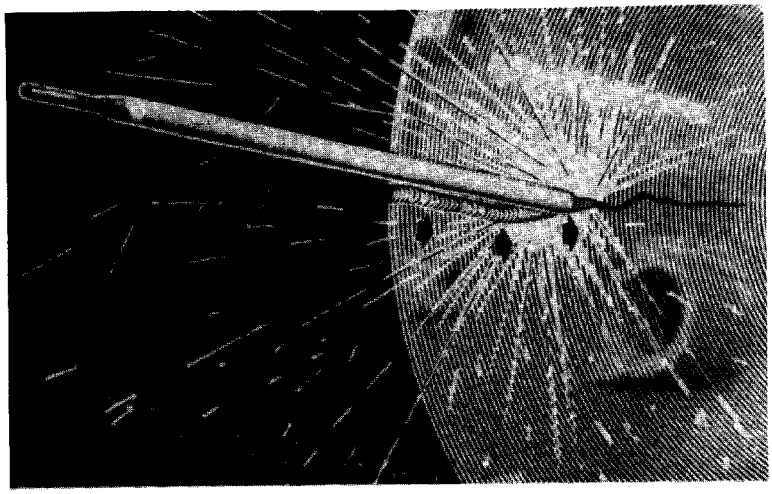


Fig. 1 Gouging out crack in cast-iron

an appreciable degree of control over the heat input, penetration, bead sequence, etc. All of these help to prevent cracks appearing in or near the welded joint.

Since damp electrodes can cause cracking in a weld and also in the heat-affected zone it is important that repair electrodes should be packed in some form of moisture-proof container or box. From this point of view, good quality plastic boxes have proved very satisfactory. In addition, they have high mechanical strength, and prevent damage to electrodes.

As already pointed out wear is not a simple material property but also depends upon such factors as lubrication, speed of movement, temperature of the environment etc. These recommendations should therefore be used as a guide, the final choice being modified in the light of practical experience.

Adhesive Wear

In situations where resistance to adhesive wear, sometimes combined with impact strength, is required a low carbon martensitic alloy steel giving a hardness of 30–35 Rockwell C is usually adequate.

Typical Applications: Hardfacing cast steel cog wheels, brake shoes, track links and rollers, crane wheels, rails etc.

The same weld metal can also be used for building up worn parts prior to hardfacing the surface. Their resistance to plastic deformation under impact and to flowing under compressive loads make the low carbon low alloy martensitic steels particularly suitable as a support for a harder but less ductile surface layer.

Recommended Electrodes for adhesive wear

	Weld metal composition %			Hardness HRC
	C	Si	Cr	
OK Selectrode 83.28	0.1	0.5	3.2	30–34
OK Selectrode 83.29	0.1	0.5	3.2	30–34

The weld metal is machinable without prior heat treatment.

Abrasive Wear

(a) low stress scratching abrasion.

For this type of application high surface hardness is essential and in many cases a martensitic low alloy steel will give good results. A hardness of 55–60 Rockwell C is desirable.

Typical applications: Ploughs, forestry equipment, loading buckets for coal, sand, soil etc., loaders, mixers, road-making machines and other objects subjected to mineral-to-metal wear.

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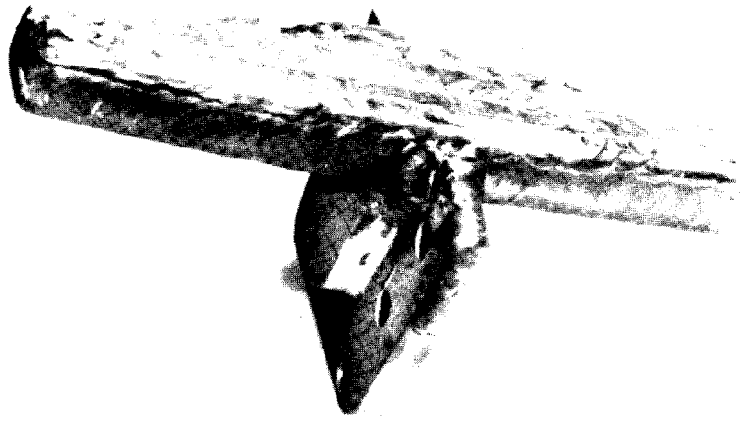


Fig. 2 Part of powder grinder hardfaced with OK Selectrode 84.78

Recommended Electrodes

Designation	Weld metal composition %					Hardness
	C	Si	Mn	Cr	Mo	HRc
OK Selectrode 83.50	0.4	0.4	0.7	6	0.6	55–58
OK Selectrode 83.65	0.7	4	0.4	2	—	58–63
OK Selectrode 84.58	0.7	0.5	0.7	10.5	—	55–58

(b) high stress abrasion

For extreme abrasion resistance such as occurs in the mineral and ore-mining industry and in brick making equipment. Details in earth-moving parts, mixer blades, excavator blades, chutes, screw conveyors and brick-making dies are typical applications.

Suitable Electrodes for high stress abrasion

	Typical Composition %		Hardness
	C	Cr	HRc
OK Selectrode 84.78 Cast or tubular rods containing a high proportion of carbides	4	33	55–60

When very high resistance to grinding abrasion is required, deposits rich in tungsten carbide are used. A typical example is OK SP 350 which consists of tungsten carbide particles in a mild steel tube. The electrode has a thin covering of flux and gives a weld metal having a hardness of 55–58 Rockwell C. The weld metal consists of tungsten carbide particles in a hard steel matrix.

Abrasive Wear with High Impact

Abrasion resistance demands hardness and impact strength demands toughness.

Since high hardness and high toughness are mutually incompatible some compromise must be accepted in practice.

Fortunately austenitic manganese steel fulfils many of the requirements. In the as-welded condition it has a hardness of about 180 Brinell, but cold works rapidly to about 45 Rockwell C.

Suitable Electrodes for abrasive wear with impact

	Weld metal composition %			Hardness
	C	Mn	Ni	As-welded
OK Selectrode 86.08	1.1	13	—	160–180 HB
OK Selectrode 86.28	0.8	14	3.5	Cold Worked 45 HRc

This type of metal is used extensively on earth moving equipment where gritty abrasion is combined with high impact. Excavator bucket lips and teeth, stone and ore crusher rolls and plates, bulldozer blades, crusher jaws etc. are all typical applications.

It should be remembered that until the weld metal has been heavily cold worked its hardness and hence its abrasion resistance is very low. To achieve initial surface hardness in the absence of heavy cold deformation, a layer of OK 83.65 or OK 84.78 can be deposited prior to putting the material in service.

Abrasive Wear with Corrosion

To combat a combination of abrasive wear and corrosion the 13% Cr martensitic stainless steels are useful. They are also oxidation resistant up to about 500°C and are thus used for hardfacing steel mill rolls where metal to metal abrasion resistance at temperatures around 500°C is required.

Resistance to combined abrasion and corrosion depends very much on the circumstances.

If the protective layer of chromium oxide is continually removed the corrosion resistance will be very much reduced.

In situations where high corrosion resistance is essential an austenitic stainless steel may give a longer service life since in addition to its corrosion resisting properties it also work hardens.

For extremely difficult conditions involving abrasion, corrosion and heat, it may be necessary to go to the cobalt and nickel based alloys.



Fig. 3 Hardfacing cone crusher with OK Selectrode 86.08

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Suggested Electrodes

Resistant to abrasion and corrosion	Weld Metal Composition % C	Si	Mn	Cr	Hardness HRc
OK Selectrode 84.58	0.7	0.5	0.7	10.5	55–58
OK Selectrode 84.42	0.12	0.5	0.3	13	44–49
OK Selectrode 84.52	0.25	0.5	0.3	13	51–56

Resistant to heat and abrasion	C	Co	W	Nb	Cr	
OK Selectrode 85.58	0.4	2	8	0.8	1.5	47–52

Resistant to high temperature oxidation, extremely resistant to wear and impact at high temperatures

	Weld Metal Composition C	Ni	Cr	Mo	Fe	W	Co	%	Hardness HRc
OK Selectrode 92.35	0.05	rest	15	16	5	4	—		40 Work Hardened
OK Selectrode 93.06	1	—	29	—	—	5	60		40–45 As-welded

OK Selectrode 92.35 gives a nickel-based weld metal which does not soften even after long periods at high temperature. The tensile strength at 800°C is above 400 N/mm² which is several times higher than iron based alloys at the same temperature. Typical applications are hardfacing hot forming tools, hot shearing tools, extrusion dies etc.

OK Selectrode 93.06 gives a cobalt based weld metal of the Stellite 6 type. It combines excellent resistance to abrasion, corrosion and oxidation in the temperature range 700–900°C. Typical applications are hardfacing of valve seatings, edges of hot forming cutting tools.

Rebuilding high speed tools

For rebuilding high speed tools a deposit which maintains its cutting edge during metal to metal wear is required. For milling cutters, turning tools, punching and stamping dies etc. a molybdenum bearing high speed tool steel is recommended.

Recommended Electrode for rebuilding high speed tools

	Weld metal composition % C	Cr	Mo	W	V	Hardness HRc
OK Selectrode 85.65	0.9	4.5	7.5	2	1.5	60–65

The weld metal has a hardness of about 60 Rockwell C in the as-deposited condition. A higher hardness and longer life is obtained if the weld metal is double tempered by heating to 525°C, holding for 1 hour, followed by air cooling and repeating the tempering once more.

Automatic and Semi-automatic Hardfacing

As can be observed in the previous paragraphs to this article, even if manual welding with coated electrodes is still the dominant process for repair and maintenance, there can be situations for the use of automatic welding processes. This is especially the case where large volumes of metal have to be deposited and when the geometrical form of the part, or parts, lends itself to automatic welding. Typical parts are for example steel mill rolls, tractor rollers, and other cylindrical parts. ESAB has developed a range of consumables for automatic hardfacing with submerged arc or with gas-shielded welding which they market under the name of Tubrodur.

These are flux-cored tubular wires made up from steel strips and then filled with flux and alloying powders. This method makes it possible to hold very strict chemical compositions, and gives better weldability than with solid alloyed wires. Alloyed solid wires with a high carbon content are difficult to draw down during production into welding wire.

Submerged Arc Welding

Submerged arc hardfacing with Tubrodur is carried out using submerged arc welding machines. Standard welding fluxes of fused or agglomerated types are used to cover the arc, and, depending on the type of alloy, these are of the neutral or alloying type. Normally, Tubrodur flux cored wires 3 or 4 mm diameter are used with welding currents of 250–500A for 3 mm and 400–600 A for 4 mm diameter Tubrodur.

Since the alloying takes place mainly from the powder core, the chemistry and hardness of the deposit will be little affected by variations in current and voltage settings. The tubular cross-section of Tubrodur gives a smoother arc with less penetration than with solid wire. This is an advantage when hardfacing because of the decreased dilution.

Submerged arc hardfacing with Tubrodur is carried out on direct current. The electrode coupled to the negative pole (i.e. straight polarity) gives a higher melting rate and less base material dilution. With the electrode connected to the positive pole (reverse polarity) the bead appearance becomes slightly better.

Gas-Shielded Metal Arc Welding

Tubrodur can also be used for semi-automatic or automatic gas-shielded metal arc welding, and can be used on nearly all standard machines which have a DC power source. Standard diameter is 2.4 mm but some of the wires are produced in the diameter 1.6 mm.

In gas-shielded arc welding the electrode is connected to the positive pole (reverse polarity) and the free wire extension (stick-out) is kept short during welding (20–40 mm).



Fig. 4
Automatic welding of a press roller

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Normal currents in this instance are about 250–300 A for 1.6 mm dia. Tubrodur and around 325–425 A for the 2.4 mm diameter.

Turbrodur wires designed for a shielding gas of CO_2 or mixed 80% Ar + 20% CO_2, however Tubrodur 15.60 can be used without any shielding gas (i.e. open arc). The tubular cross-section is combined with the flux constituents in the core and gives a smooth arc with low penetration when compared with solid wire welding.

Filler material for:

Building-up
OK Tubrodur 15.40 HRC 35 C 0.1 – Cr 3.5
OK Tubrodur 15.42 HRC 42 C 0.2 – Cr 4.5 – Ni 0.5 – Mo 0.5

These wires are used for the rebuilding of carbon and low alloy steels used on tractor rollers, crane wheels, shafts, gears. They also give an excellent base for hardfacing with harder weld metals. The deposit is easily machineable.

For submerged arc welding ESAB OK Flux 10.40 or OK Flux 10.80 and 3 or 4 mm dia. wire or 4 mm wire should be used. With OK Flux 10.80 hardness will be somewhat higher.

For gas-shielded metal arc welding use CO_2-gas and a 2.4 mm dia. Tubrodur.

High Impact with Moderate Abrasion
OK Tubrodur 15.60 HV 250 work hardness to 45 HRC C 1 – Mn 13

This wire is recommended for rebuilding of worn parts of cast 13% manganese steels, such as hammer crushers, jaw crushers, shovels, dredger pumps. The austenitic weld metal is very tough and impact resistant. Here the wire is available in 1.6 and 2.4 mm dia. for gas metal arc welding with CO_2 or for 'open-arc' welding.

High Abrasion with Moderate Impact
OK Tubrodur 15.50 HRC 55 C 0.3 – Cr 5 – Mo 1.0
OK Tubrodur 15.83 HRC 55 C 0.7 – Cr 4.5 – Mo 3 – W 4

These wires are used for hardfacing of carbon and low alloy steel parts, subject to abrasive wear. Mixers, feeder screws, dredger pump parts, etc. are typical examples.

It is recommended in these cases to preheat the part to be welded to about 200°C and to allow slow cooling. N.B. The deposit is not machinable.

For submerged arc welding use OK Flux 10.40 with OK Tubrodur 15.50 and OK Flux 10.61 with OK Tubrodur 15.83 in 3 or 4 mm dia.

For gas-shielded metal arc welding CO_2 is used as the shielding gas and 2.4 mm dia. wire. OK Tubrodur 15.50 is also available in 1.6 mm dia.

Friction and Corrosion
OK Tubrodur 15.70* HRC 32 C 0.06 – Cr 13
OK Tubrodur 15.73 HRC 50 C 0.3 – Cr 13
OK Tubrodur 15.74* HRC 50 C 0.3 – Cr 13

These wires are used for surfacing and hardfacing of parts subject to both corrosion and wear, even at elevated temperatures. Examples are valve seats, pump parts, etc.

The deposit of OK Tubrodur 15.70 can be machined as-welded. OK Tubrodur 15.73 and OK Tubrodur 15.74 both give a weld deposit with the same analysis but OK Tubrodur 15.73 is designed to operate at its best on positive electrode polarity using submerged-arc with OK Flux 10.91 and also CO_2-welding. Note that OK Tubrodur 15.74 is designed to operate at its best on negative electrode polarity.

*only for submerged arc welding with OK Flux 10.91.

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2. WELDING OF CAST IRON

The welding of grey cast iron and ductile iron is not as simple as the welding of ordinary steel, but tough crack-free welds can be obtained when using the correct procedure and consumables.

The welding of cast iron covers the following:

(a) Constructional welding of cast iron to itself or to steel
(b) Repair welding of casting defects
(c) Repair welding of damaged or worn parts

Main Cast Iron Types
(a) Grey cast iron (flake graphite)
(b) Ductile iron (spheroidal graphite)
(c) White iron
(d) Malleable iron
(e) Alloyed cast iron.

The following table indicates typical chemistry and mechanical properties for these types of cast iron compared to cast steel.

		Ductile iron	Grey cast iron	Malleable iron	Cast steel
Chemical Analysis %	C	3.3–3.8	2.8–3.8	2.2–3.4	0.2–0.4
	Si	1.5–2.8	1.7–2.8	0.4–1.6	0.3–0.5
	Mn	0.1–0.3	0.2–0.8	0.2–0.5	0.45–1.2
	P	≤0.01	<0.3	0.07–0.25	0.035
	S	≤0.01	<0.12	≤0.1	0.035
	Mg	0.03–0.08	—	—	—
Brinell Hardness HB (kp/mm²)		130–320	150–280	150–270	150–320
Tensile Strength (kp/mm²)		35– 70	25	35– 70	38– 70
Yield Strength (kp/mm²)		22– 50	—	20– 55	19– 42
Elongation δ_5%		2– 25	—	2– 12	12– 25

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All cast irons are characterized by an excess carbon content (3–4%) which while solidifying separates in the matrix metal and, depending upon the shape and form of this carbon, the microstructure. Thus the cast iron properties are determined.

Grey cast iron has graphite in the form of flakes. *Fig. 5.*

By using special inoculants, such as Mg, the graphite can be obtained as spheres. This cast iron is referred to as *ductile iron* or alternatively spheroidal graphite iron. *Fig. 6.*

Rapid cooling will not permit the formation of graphite. Instead hard and brittle iron carbide (or cementite) is obtained, thus giving a *white iron.*

High alloy irons contain additions of nickel, chromium, molybdenum, etc. to increase corrosion or wear resistance.

Examples are Ni-hard with about 5% Ni and Ni-Resist with 10–30% Ni and 1–5% Cr.

The matrix (surrounding the graphite) properties can be treated to give a soft ferritic structure or a harder *perlitic* structure.

Sulphur and phosphorous are very low in ductile iron but grey cast iron may contain between 0.1 and 1.0% P to improve fluidity and the casting of thin walled parts or to increase hardness and consequently wear resistance.

Cold Welding of Cast Iron

Cold welding is the name for the arc welding of cast irons using nickel-based welding electrodes. The nickel deposit gives a ductile machinable weld even when welding at ambient temperature. Because of its simplicity this is the widespread method for the welding of cast iron. During the welding process part of the cast iron material will be melted and mixed with the deposited electrode metal. This part is referred to as the *fusion zone.* The nickel weld metal can dissolve high levels of carbon and iron from the base material without becoming hard and brittle even under rapid cooling conditions. The nickel base weld metal also has a coefficient of thermal expansion very close to that of cast iron, which is a further advantage in reducing the welding stresses. If normal mild steel weld metal was used, it would become extremely hard and brittle because the carbon

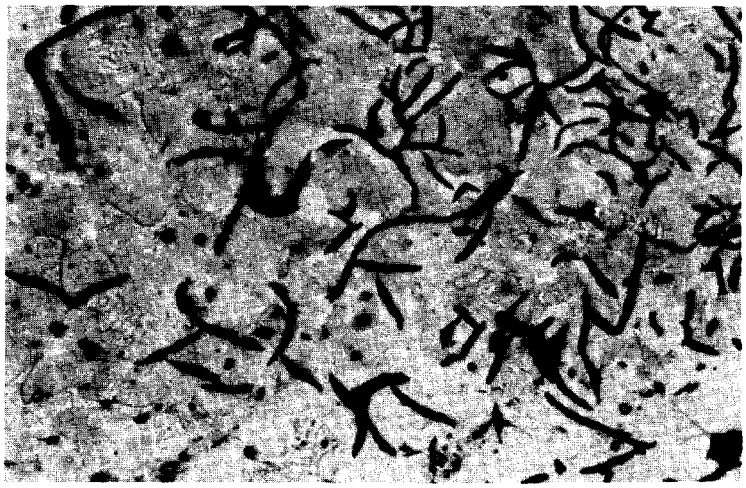


Fig. 5 Grey cast-iron (flake graphite)

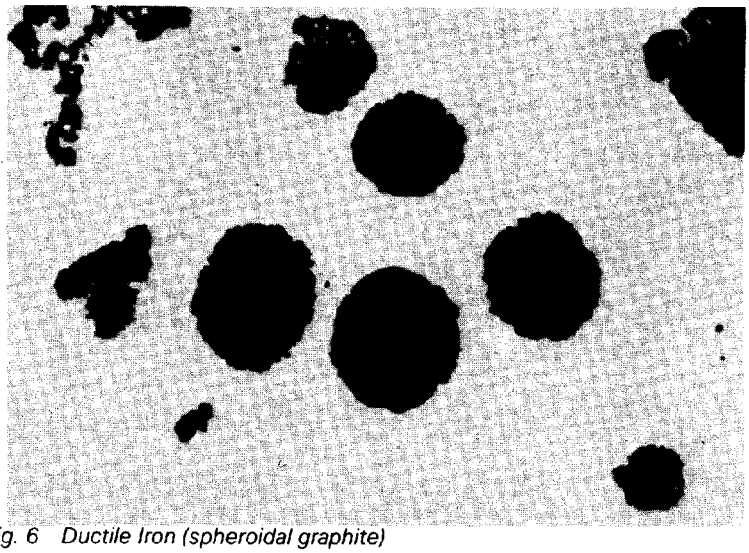


Fig. 6 Ductile Iron (spheroidal graphite)

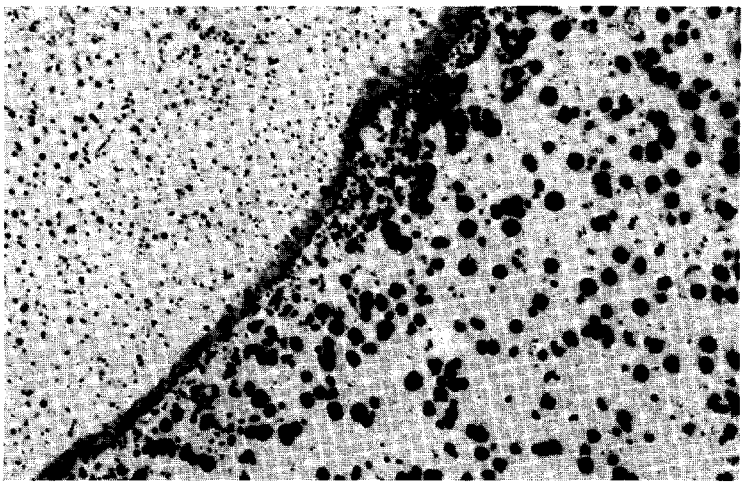


Fig. 7 (a) Micrograph OK Selectrode 92.58 weld metal Ductile Iron base metal.

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content from the dilution would result in martensite and iron carbide formations during solidification.

The *partial fused* and *heat affected zone* is the part in the base material adjacent to the fusion zone where the heating has resulted in partial melting or at least in transformation of the micro-structure. It is here that the highest hardness in a weld in cast iron will be found, since the rapid cooling cycle under welding will not allow graphite formation but instead martensite and white iron will be formed.

The heat affected zone hardness is a function of the base material composition and the heating and cooling conditions. A ferritic base cast iron will result in lower hardness than a perlitic grade.

The principle behind cold welding is to reduce the amount of hard constituents and to minimize the width of the excessively hard zone by using a low heat input. *Fig. 7.* Even *preheating* will not stop the formation of a hard zone but may reduce the peak hardness because of a decreased cooling rate. This can thus improve machinability after welding.

In multipass welds the heat from subsequent runs on top of the previous welds can be utilized to anneal and soften the heat affected zones.

Cold Welding Nickel Electrodes

Special nickel-based electrodes with a basic-graphitic flux coating on a nickel-type core wire have been developed for cast iron welding.

These electrodes have been designed to weld on low currents and can be welded with both AC and DC, in all positions. The chemistry for alloying additions and trace elements is composed to give a ductile weld metal able to give sound crack-free welds.

Standard Electrodes:

	Main composition %	Classification	UTS N/mm²	HV
OK Selectrode 92.18	C = 1.0 Ni = 95	AWS ENi-CI DIN 8573 ENiG2	390	170
OK Selectrode 92.58	C = 0.7 Ni = 55 FE = 43	AWS ENiFe-CI DIN 8573 ENiFeG2	550	190
OK Selectrode 92.78	C = 0.5 Ni = 66 Cu = 36	AWS ENiCu-B DIN 8573 ENiCuG2	450	180

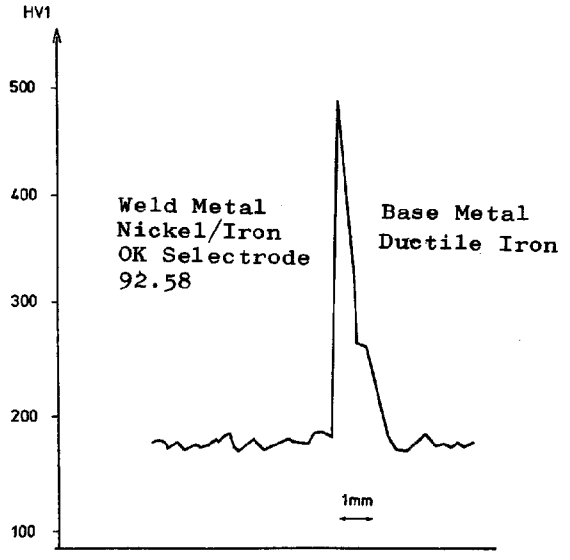


Fig. 7 (b) Micro-hardness (HV, 1 Kg load) across metal

The softest and consequently the easiest machinable weld metal is obtained by using the almost pure nickel electrode OK Selectrode 92.18. OK Selectrode 92.18 should therefore be used where this property is considered to be the most important. The tensile strength of weld metal may be too low for some ductile iron materials. The nickel-iron electrode OK Selectrode 92.58 is then recommended for materials when high strength is required.

OK Selectrode 92.58 may also be used for welding of grey cast iron. Due to the higher hot crack resistance of the nickel/iron weld metal, it is recommended for phosphorous containing grades.

The copper/nickel weld metal of OK Selectrode 92.78 gives an intermediate strength between what is obtained with the former electrodes. Generally the colour matching with the cast iron is best when it is weld metal from OK Selectrode 92.78.

The chemical corrosion resistance of nickel/copper weld metal may also be of use with high alloy austenitic castings.

Welding Procedure

In order to obtain a satisfactory weld, the work-piece must be carefully prepared before welding.

Grey cast iron which has been used for a long time can often be impregnated with oil or grease and must be cleaned before welding. The most efficient way is to use a high pressure steam cleaning operation. The casting should also be cleaned in a solvent.

Castings that have been used at very high temperatures and become 'burnt' should not be welded; this is also true for heavily steam-oxidized cast-iron.

Prior to welding on new castings the casting skin and sand inclusions must be removed from the surface to be joined especially in the area of welding.

Butt joints in cast iron are wider than those for mild steel. The angle is 60–75° for V-joints and 20–30° for U-joints.

Whenever possible in heavy sections a double V-joint or double U-joint is preferred to minimize distortion and welding stresses.

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It is essential to remove all sharp corners to avoid local spots of high dilution when welding.

Joints can be prepared by machining, grinding or arc-gouging. The gouging electrode OK Selectrode 21.03 is very efficient in joint preparation. *See Fig. 1.* In particular in gouging out cracks and defects. Care must be taken to remove any remaining slag after the gouging operation.

Before making a weld repair the damaged area should be checked with a penetrant dye to locate the extent of possible small cracks which may be invisible to the naked eye.

To prevent cracks from propagating further it is recommended that a hole be drilled at the end of the crack before welding is begun.

The proper welding electrode has been selected according to the requirements of the weld and with regard to the type of cast iron to be welded. If possible the workpiece should always be positioned so that welding can be carried out in the flat position.

The first welding passes are most critical and should be made with a 2.5 mm diameter electrode or for heavier sections with a 3.25 mm diameter electrode; the lowest possible current should be used. This means that dilution of the base material and the size of the heat affected zone will be minimized.

Generally only stringer beads should be applied; however a slight weaving (less than twice the electrode diameter) can be used to obtain good wash at the sides of a joint. In order to reduce stresses and heat input all weld passes should be made in short runs of 25 mm max. and preferably shorter on thin material.

Immediately after deposition, all runs should be peened with a ball pointed hammer to deform the bead and obtain compression stresses which are preferable to tension stresses on the weld.

After the sides have been surfaced, the center should be gradually filled in. These latter passes will anneal the heat affected zone.

Between each pass all slag residues must be carefully removed. It is very important to fill in the crater end before the arc is extinguished. This is done by a slight back move of the electrode. Whenever possible the arc should be struck on top of a previous run and the arc must never be directed at a corner or an edge.

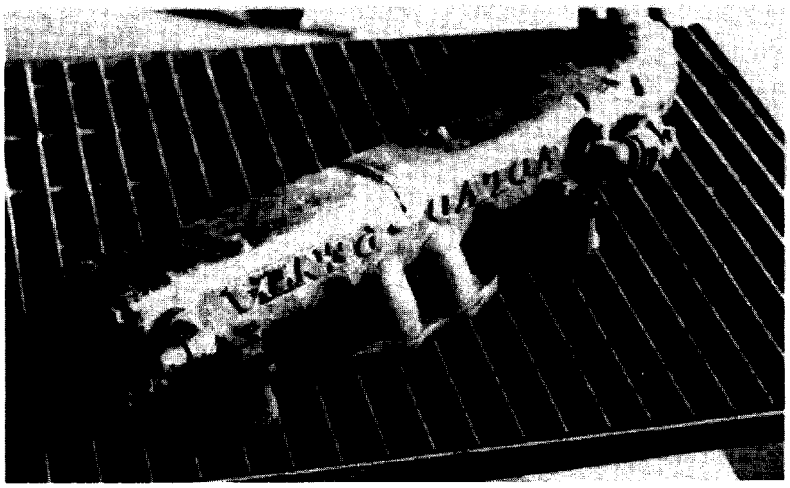


Fig. 8 Repaired exhaust manifold on Volvo-Penta motor

Welding with a Steel Electrode

Mild steel is the least expensive filler material and special mild steel electrodes may be used for some cast iron welding applications.

Electrode	Weld metal analysis	Classification
OK Selectrode 91.58	Mild steel	AWS Est DIN 8573 EFeS2

The weld metal will pick up carbon from the cast iron and become hardened an unmachinable.

The use of a steel electrode is therefore limited to the repair of smaller defects and to welds where there is less demand for weld metal toughness.

After using OK Selectrode 91.58, welds can be chromium plated or enamelled and have excellent colour which matches with the cast iron. This electrode also may be used on dirty oil impregnated cast iron and on burnt or oxidized castings, due to its special flux coating. *Fig. 8.*

3. WELDING DIFFICULT STEELS

When making repairs one is often faced with the problem of welding steels which are either of unknown composition or which have a composition which makes them difficult to weld without cracking. They are often hardenable cast steels which have a high carbon content or a high sulphur content. Examples are free cutting steels in shafts and worn parts which have to be welded to carbon steels. A great many solutions have been put forward to solve problems such as these, e.g. the use of extra hard dried basic covered electrodes or overalloyed' stainless electrodes. During recent years interest has been concentrated on electrodes giving a 29% Chromium/9% Nickel weld metal whose composition ensures that a very crack resistant weld metal is formed even when diluted with high carbon steel.

The stainless weld metal holds any hydrogen in solution thus minimizing the risk of cold cracking in the weld and heat affected zone. As is in the case of welding of cast iron (see paragraphs which follow) dilution should be kept to a minimum and heat input kept low. Because of this, the electrode diameters selected are small compared with those used on mild steel of the same thickness when production welding. Welding current is kept low and a planned bead sequence is used (N.B. Weaving should normally be avoided since it increases heat input).

Examples for the use of this type of electrode is the repair of cracks in steel castings and the building up of worn shafts. When hardfacing with electrodes giving a very hard deposit, a buffer layer deposited with a 29 Cr-9 Ni electrode is recommended. This evens out the welding stresses. Because of its high chromium content the weld metal has very good corrosion resistance and a yield stress in excess of 70 kg/mm².

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Electrodes for Welding Difficult Steels

	Hardness	Weld Composition	Properties – Applications
OK Selectrode 67.45 OK Selectrode 67.52	190 HV 190 HV	Cr = 18 Ni = 9 Mn = 6 Cr = 18 Ni = 9 Mn = 6	For butt welding austenitic manganese steel and for stainless cladding. Tensile strength 690 N/mm².
OK Selectrode 68.81	240 HV	Cr = 29 Ni = 9	For welding 'difficult' steels. Tensile strength 800 N/mm².

4. WELDING ALUMINIUM

Cast aluminium alloys such as Silumin which contain about 11% silicon are used for gearbox housings and cylinder heads and for parts where it is necessary to keep the weight as low as possible. The aluminium has generally been rolled or extruded. For many years gas welding and arc welding with covered electrodes were the processes for welding aluminium and its alloys, but today inert gas shielded methods, such as MIG and TIG, are dominant in this field. In spite of this there is still a place for welding with covered electrodes because of its particular advantages for on-site welding along with the inexpensive and simple handwelding equipment required. If aluminium is infrequent, it is not economic to purchase expensive semi-automatic or automatic machines. Also converting existing equipment used for the welding of steels with CO_2 gas is also hardly worthwhile just for an occasional weld. Cheaper is covered aluminium electrodes which can be used in all positions but a DC power source is then essential. Welding speed is relatively high and accuracy in joint preparation is not particularly important. Here the formation of oxide films is the most difficult problem to overcome and the joints should, as far as it is practicable, be welded in one run. Preheat of the parent plate makes welding easier and leads to better quality weld metal. Two main types of electrodes are available, one containing 11% silicon which welds most of the cast alloys and a large number of the wrought types; the other which contains 1.25% manganese is for welding rolled plates.

Electrodes for Welding Aluminium

	Weld Metal Composition	Tensile Strength N/mm²	Properties – Applications
OK Selectrode 96.20	Mn = 1.3 Rest. Al	100	General purpose of electrode for welding wrought Al, AlMn and AlMgSi alloys.
OK Selectrode 96.50	Si = 11 Rest. Al	170	For welding cast aluminium alloy (Silumin) and certain rolled alloys.

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Repairing Cast Iron by Welding

. . . it isn't easy, but knowing how puts the odds in your favor

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CAST IRON IS A TERM applied to a number of iron compositions distinguished from other iron alloys by their high carbon content and from each other primarily by the pattern of distribution of carbon in the microstructure of each. The carbon content of cast irons varies from 1.7 to 4.5 percent, but the most common types fall within the 2.5 to 3.5-percent range. The metallurgical structure of the alloys is more significant in determining their mechanical properties than the absolute level of carbon in their compositions.

Types of Cast Iron—All alloys lumped under this name are difficult to weld because of their high carbon content. There is a tendency for massive carbide deposits to form in the area adjacent to the weld, and for high-carbon martensite to form in the remainder of the heat-affected zone. Both these materials are brittle and may crack under load or spontaneously. Nevertheless, it is useful for the plant engineer to be able to distinguish between the various alloys.

White cast iron maintains its carbon in combined form. Its name derives from the silvery-white appearance of a fractured surface—an appearance that results from the absence of graphitic carbon. It is extremely hard (up to 600 BHN); but because it is also brittle, it is not widely used.

Malleable cast iron is the result of heat treatment that disperses the carbon as tempered carbon in a ferrite matrix. It is more ductile than white cast iron and has greater tensile strength. In castings, the surface structure differs from the core.

Nodular cast iron, which is similar to malleable cast iron in its mechanical properties, is produced by adding alloying elements that cause the carbon to disperse in tiny balls, or nodules. It is also called *ductile iron*.

Gray cast iron—so called because the carbon flakes that cover its fractured surfaces give them a gray appearance—is the most common of the cast irons. It is the addition of 1 to 3 percent silicon, combined with slow cooling, that produces the characteristic precipitation of carbon as flakes of graphite in the microstructure. Welding gray cast iron may be complicated by the presence of alloying elements, such as nickel.

Preparations for Welding—Although the results from welding cast iron are somewhat unpredictable, experience can greatly reduce the failure rate. The plant engineer should first consider whether the casting is, in fact, repairable and worth repairing, keeping in mind the type of cast iron involved, the service conditions the casting must endure, the severity of contamination of the weld area, the size of the casting, the wall thickness, and whether the welds will have to be made in difficult positions.

If the decision is favorable, the plant engineer must decide on the cleaning procedure, welding method, type of electrode, and joint design, and must determine the need for preheating.

The welding method chosen will depend in part on the alloy to be welded. Cast iron can be joined by the oxyacetylene, gas metal-arc (GMAW), and shielded metal-arc (SMAW) processes. Oxyacetylene braze welding with bronze rods is the preferred process for certain malleable casting repairs; GMAW is considered more suitable for seam welds in malleable and ductile irons than for repair of gray iron castings.

SMAW is probably the best repair process for most purposes. It permits all-position welding, sometimes eliminates the need for preheating, and may permit a better color match than other processes. It is also widely known and used, and may therefore be the only method available.

Preliminary treatment of the casting is much the same for all the alloys. The graphite "skin" and all surface contamination must be eliminated by sanding, grinding, or some other means. Grease and oil must be totally removed, and all unsound metal must be ground away from the joint area.

Oil that has seeped into the casting is a common contaminant. An easy method for removing it is to heat the entire casting to 700 or 800 F for 30 or 40 minutes. Either an oven or a torch can be used. If the heating is done with a torch, the part should be brushed occasionally and the process continued until no more oil appears on the surface.

Porous welds on a clean casting usually indicate that the cast metal contains an excess of trapped hydrogen. It can usually be eliminated by heating the casting to a dull red for a few minutes. In some cases it will not be possible to heat the casting before welding; a sound repair can still be made by running a bead, grinding away the porous areas, and repeating the procedure until a good weld is achieved.

Joint design for cast iron generally calls for wider bevels than are needed for joints in steel. Bevels can be cut by grinding, chipping, and arc-air gouging, or made with SMAW grooving electrodes; but if either arc method is used, the oxidized metal must be ground away.

A single V groove, Fig. 1, is usually adequate for metal up to ⅜ in. thick. The bevels on each side of the joint should be 40 degrees or a little greater, for an included angle of 80 to 90 degrees. The root face should be approximately $\frac{1}{16}$ in.

Thicker castings require a double-V groove, Fig. 2, with the same bevel and root-face specifications. The

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"Preheating is unnecessary unless the casting is crack sensitive."

surface should always be ground at least ½ in. back from the bevel. If the section is very thick, but can be grooved on only one side, a modified U joint will reduce the amount of filler metal needed.

The best way to ensure a full-penetration weld is to grind the underside of the bead and run another pass on the back. If full penetration at the root is essential, but only one side of the joint is accessible, it is possible to grind the joint with a 1/16 to ⅛-in. root gap; this practice ensures full penetration, but alignment is much more difficult. Jigging or clamping the parts is often necessary, and strong tack welds at each end of the joint are a good precaution.

Coated electrodes of several kinds are available for joining cast iron. The optimum choice will depend on the material and condition of the casting, and the service required of the part. Soft steel electrodes with special low-current flux coatings provide an economical option. Welds made with these electrodes are not machinable, but have better than average strength. Soft steel electrodes are especially useful for repairing engine blocks, grounds-maintenance implements, and similar items. They are also useful for making tie-in layers for replacing broken gear teeth, and in other repair operations in which it is necessary to build up an area with an electrode normally used for mild steel.

A 100-percent-nickel electrode is useful for joining thin, hard-to-weld sections when machinability is nec-

ELECTRODE SIZES FOR REPAIR WELDS IN CAST IRON

Section thickness, in.	Electrode size, in.
0 to 3/16	3/32
3/16 to 3/4	1/8
3/4 to 1 1/2	5/32
1 1/2 or thicker	3/16

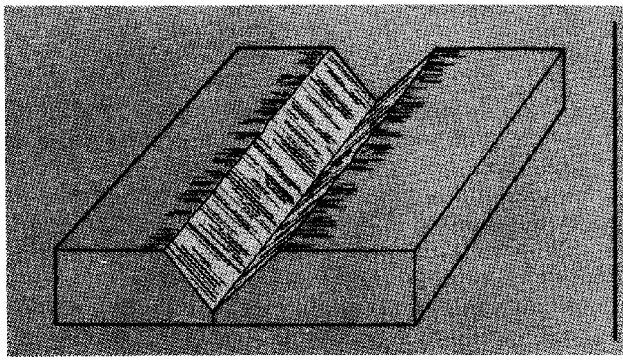


Fig. 1. A simple V groove is usually satisfactory for welding cast iron no thicker than ⅜ in. The bottom of the V should be about 1/16 in. from the lower surface of the stock.

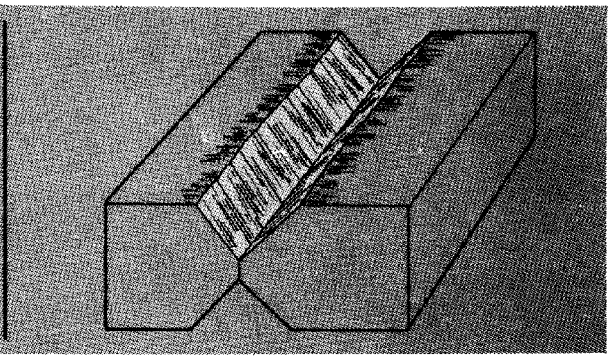


Fig. 2. A double-V groove is usually desirable when thick castings must be welded, because it requires the deposition of less weld metal.

Fig. 3. Skillful preheating not only opens the fracture for better welding, but also puts the weld in compression when the workpiece has cooled.

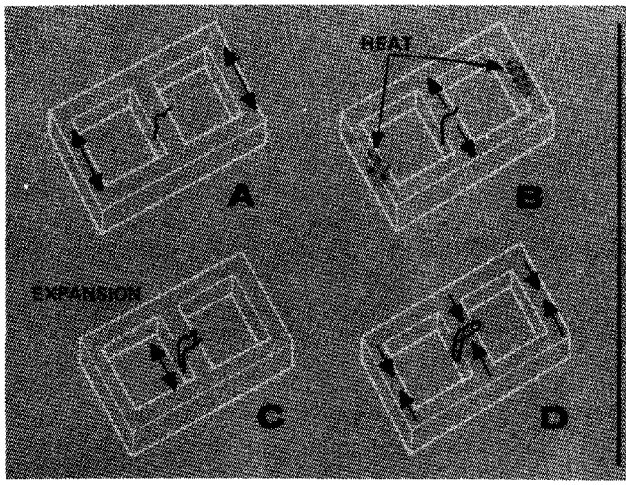
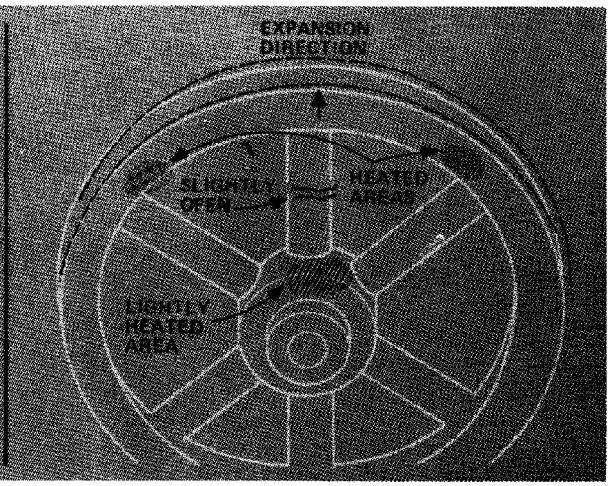


Fig. 4. Cracks in the spokes of handwheels, flywheels, and gears can be opened by heating the hub and the rim to the left and right of the damaged spoke.



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essary but high strength is not. The weld metal is soft, ductile, and, when properly applied, free of porosity. The arc is very smooth. Preheating is unnecessary unless the casting is exceptionally crack sensitive. Cylinder heads and pump housings are good examples of the type of part that can be repaired with these electrodes.

Nickel-iron electrodes are usually preferred for heavy sections and applications requiring high strength. They are useful for repairing very badly cracked engine blocks and cylinder heads, heavy-duty pump housings, and high-phosphorus, ductile-iron, and nickel-alloy cast-iron castings. Machining the welds is difficult but not impossible.

Low-hydrogen electrodes, although not specifically intended for welding cast iron, are useful for repairing new castings when color match and economy are important. They can also be employed to repair broken castings, provided the welding can be done in the flat position. E7018 electrodes are the most commonly used type. All low-hydrogen rods must be kept absolutely dry, or their effectiveness will be greatly reduced.

Making the Weld—Alternating current and direct current with the electrode positive give the best results on cast iron. The electrode diameter should be the same as that used for a similar joint in steel, or one size smaller. (The smaller size helps hold the heat input down, which can be especially helpful when small or thin castings must be welded.) Although almost all thicknesses could, in a pinch, be welded with a ⅛-in. electrode, the sizes listed in the table will give optimum results.

Welding current should be held at the lowest level that will produce a smooth arc and an even flow of molten metal. Alternately, a higher current can be used with a faster-than-normal travel rate. Slightly more current is needed for a-c than for d-c electrodes of the same size, and steel and nickel-steel electrodes require somewhat higher currents than do 100-percent-nickel rods. Preheated castings will take somewhat less current than cold ones. When possible, test welds should be made on a piece of scrap cast iron similar to the piece to be welded to establish the correct current and the best welding technique.

Preheating provides a number of benefits, although certain electrodes and procedures do not demand it. These benefits include good amalgamation of the base metal with the filler metal, prevention of the formation of chilled metal at the line of fusion, accommodation of expansion in the casting, and the elimination of gas pockets. Thus, despite its being an extra step in the repair process, judicious preheating can be used to advantage when it is necessary to produce a strong, ductile, and machinable weld joint. It is axiomatic that castings that are known to be crack sensitive should be preheated.

Unless the casting is very large, the usual practice is to heat the entire piece to between 500 and 600 F and cover it except for the area to be welded. If it is known that the casting is exceptionally hard or has considerable internal stress, preheating to between 800 and 900 F is indicated. After being welded, the casting should remain covered until it is cool.

Localized heating of various parts of the casting is a valuable technique for neutralizing areas of special stress. It can also be used to expand areas stressed by heat in the weld area, and careful application of heat can put the weld under a slight compressive stress when it cools, Fig. 3.

Cast gears and flywheels present special problems; the procedure should be carefully planned before a torch is put to the metal. It is usually better not to preheat the entire casting unless it is without spokes—otherwise, unequal stresses may develop on cooling.

In flywheels with spokes (a very common design), the fracture usually occurs in one or more spokes. Moderate heat at the hub and at two points on the rim, Fig. 4, will open the crack slightly, making it easier to weld. If the rim itself is cracked, heating the spokes on each side of the break will have a similar effect.

Good welding practice forbids trying to infuse too much weld metal at one time. Weaving should be avoided. Techniques that limit heat input—such as the use of stringer beads—should be used. The metal adjacent to the weld area should never be permitted to become cherry red. Beads should be kept short—as short as 3 in. on some small castings. Larger castings can tolerate greater bead length, and the experienced welder soon learns to judge the maximum bead length he can use on a particular job.

The skip technique is used on long joints to minimize stress buildup and distortion. A bead is run at each end of the joint, then in the middle, after which the space between these beads is filled with short beads placed alternately at the left and right ends of the joint. If the joint requires more than one pass, the welder should continue the same procedure, taking care not to break the arc over a spot where it was broken previously.

The backstep technique is especially useful for welding shallow V joints. The beads are run in the opposite direction to the weld so that each new bead ends at the point where the previous bead began.

Travel should be at a speed that deposits a bead about two electrode diameters wide, no matter which technique is used. To prevent hairline cracks, fill the crater before breaking the arc.

Slag removal should proceed from the starting point of the bead to the stopping point, to avoid hammering slag into soft metal. When the bead is clear of slag, peening can be begun at the point where the arc was broken. It should proceed toward the starting point. Only the bead should be peened, and excessive peening should be avoided—10 to 15 blows an inch are sufficient. Peening the sides of the joint will increase the hardness of the metal in the heat-affected zone.

Temperature should be checked after each pass by touching the weld area with the bare hand. If the metal is too hot for prolonged contact, additional cooling time should be allowed before the next pass is begun. (Con't.)

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"The preparation of a malleable casting for welding is the same as the preparation of a gray-iron one."

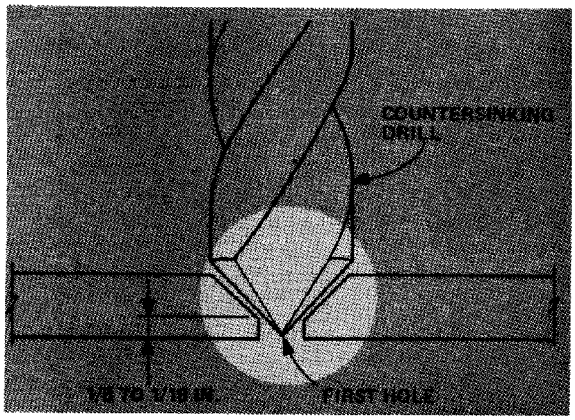


Fig. 5. Countersunk holes at the ends of a block crack prevent later stress concentration there. A ¼-in. drill should be used to make a pilot hole.

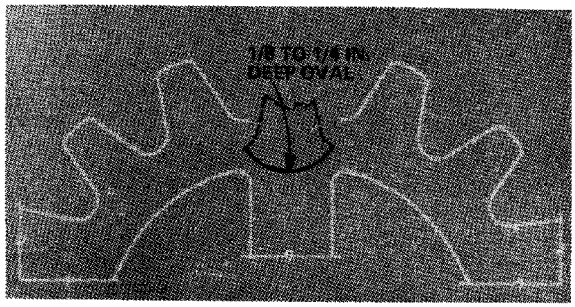
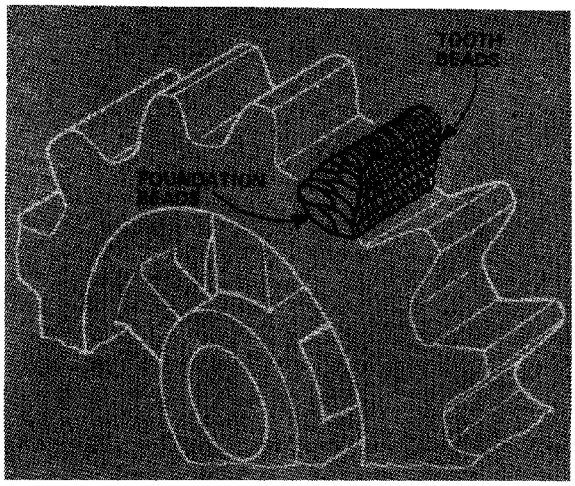


Fig. 6. Badly damaged cast-iron gear teeth should be replaced, not repaired. The rim below the old tooth's root must be deeply ground to provide a good foundation for the new one.

Fig. 7. The foundation beads of the new tooth should be laid with a high-strength rod. The beads from which it will be machined can be of a more easily worked alloy.



Malleable castings offer a special challenge because of their heat sensitivity. Their malleable nature results from long annealing under controlled conditions, not from their metallurgy; and welding heat can return the metal in the weld zone to its original white iron state. Oxyacetylene braze welding with bronze rods avoids the difficulty, but cannot always be used. Arc braze welding with coated phosphor-bronze electrodes (or with specially formulated steel electrodes) is a convenient alternative, but the nature of the arc process makes some melting of the base metal unavoidable. If it is essential that the metal in the weld area retain its malleable characteristics, the heat treatment that produced them originally must be repeated.

The preparation of a malleable casting for welding is the same as the preparation of a gray-iron one, but it is even more important that the casting be kept cool during welding. The arc should never be permitted to bring the base metal to red heat.

Examples—An examination of some common repair jobs will show how the principles described in this article are applied.

Welding cracked engine blocks is a typical maintenance shop assignment. (The procedure is the same for cylinder heads.) The first step is to clean the area around the crack thoroughly. Next, the casting should be heated just enough to open the crack slightly. A punch mark should be made at the ends of the crack, or a very short distance beyond them, to facilitate starting a ¼-in. drill. The hole should go completely through the metal. A considerably larger drill should then be used to countersink the holes to within ⅛ in. of the bottom, Fig. 5. Next, the crack should be V'd out with a chisel or a grinding disc. The bottom of the V should be at the same depth as the bottoms of the countersunk holes. The crack can then be welded using a soft steel electrode and the backstep technique.

Replacing broken gear teeth is another common repair. The first step is the complete removal of the broken tooth, and the grinding of the base area to a deep oval, Fig. 6. Foundation beads should be deposited with either a low-hydrogen electrode or a mild-steel electrode with a special coating for welding cast iron. The use of these electrodes will ensure a strong foundation for the new tooth. After the foundation is in place, the metal from which the new tooth will be machined can be deposited with an ordinary mild steel electrode, Fig. 7.

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THE TECHNOLOGICAL EFFECTIVENESS OF METHODS OF WELDING CAST IRON

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VNIILITMASh

SYNOPSIS

Welding with preheating occupies a leading position in the repair of iron castings. In the repair of components worn in service the principal requirements usually concern the strength properties. Welding without preheating can be used in these cases.

The large volume of welding jobs in the repair, manufacture and service of iron castings and components requires a continuous improvement of welding methods and welding consumables used for this purpose. As a result of demands from industry, a large number of scientific and research studies have been carried out to find highly efficient methods for the manual and mechanised welding and surfaces of cast iron. Improved welding consumables have also been developed.

At the current level of the technology of welding cast iron, welding operations involved in the repair of defects include general or partial preheating to 600-650°C (hot welding) - cast iron forms in the deposit (joint); with low-temperature preheating to 150-400°C or without preheating (cold welding) - this technique results in the formation of alloys (not cast iron) in the deposit (joint), which can be machined by cutting.

Welding and surfacing processes are manual or mechanised, and they can be carried out with melting of the parent metal (welding) or with the formation of a large volume pool.

The most efficient methods or repairing defects are given in the Table, together with their application areas. The hot arc mechanised welding of cast iron with a flux-cored wire using specialised single- and three-wire semi-automatic equipment, and also semi-automatic welding with ceramic rods, are the most progressive methods of welding cast iron. In addition, the application of special heating furnaces and devices for the preheating of small, medium and large components made it possible to increase considerably the welding and technological parameters of the hot welding processes.

The hot gas welding of cast iron with natural (supply line) gas, using single-flame nozzles with cumulative action, and also multi-flame wire gauze nozzles, replaced acetylene welding, thus reducing welding costs.

Low-temperature braze welding (without melting of the parent metal) with cast iron, brass brazing alloys and powdered alloys (type NPCh), and also methods of gas powder surfacing, make it possible to solve efficiently the problems associated with the repair of defects on machined surfaces of components.

A significant contribution to the increase in the cold welding processes for cast iron has been the development of a wide range of electrodes based on iron, nickel and copper, and alloys of these metals. These electrodes include thin, self-fluxing, nickel- and copper-base wires; and the semi-automatic welding of thin-walled cast iron components with these consumable is characterised by high technological parameters.

Thus, by selecting welding methods and corresponding welding consumables it is possible to produce various deposited metals, e.g. cast iron with the required properties, or alloys, which correspond with the working features of components. On the basis of the properties of deposited metal, all the welding of cast iron can be divided into two groups. The first group includes the deposited metal, consisting of cast iron; the second group is represented by the deposited metal, consisting of various alloys with good

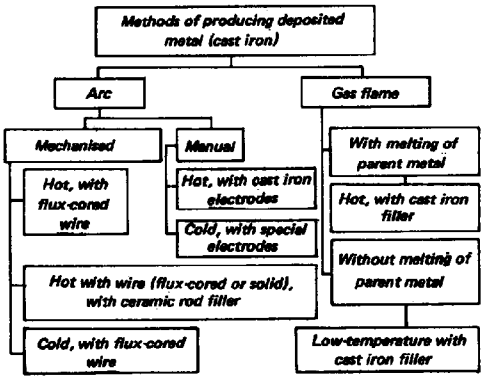


Fig.1. Classification of welding methods producing cast iron in deposited metal.

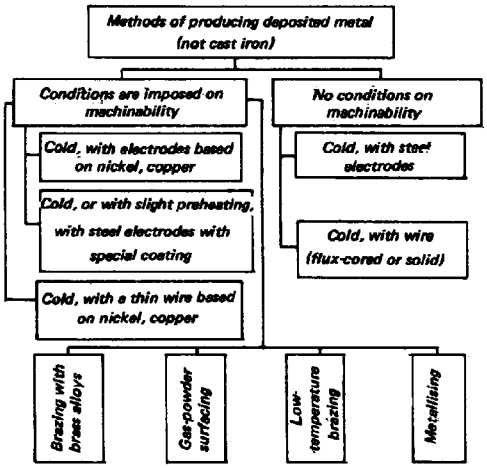


Fig.2. Classification of welding methods producing various alloys in deposited metal.

Source: Svarochnoe Prozvodstvo (Welding Production), 213(11), 1976 (Translation published by the British Library)

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Treatment of cast iron	Characteristics of the defect to be repaired	Requirements on deposited metal and welded joint (the type of deposited metal
Deposited metal (cast iron) with prescribed properties*		
Hot gas welding, with cast iron filler	Small and medium-sized defects on machined surfaces of castings subjected, or not subjected, to surface hardening	Good machinability and density, similarity with parent metal in structure, hardness, strength and colour
Gas low-temperature braze welding, with cast iron filler	Small blind defects on machined working surfaces	Good machinability and density, similarity with parent metal in structure, hardness, strength and colour
Hot arc welding, the pool technique: manual with cast iron electrodes mechanised with flux-cored wire	Large defects on surfaces being machined and on surfaces after machining; surfaces subjected, and not subjected, to hardening	Good machinability and density, similarity with parent metal in structure, hardness, strength and colour
mechanised, with flux-cored wire with ceramic rod	Small and medium-sized defects on machined surfaces not subjected to surface hardening	Good machinability and density, similarity with parent metal in structure, hardness, strength and colour
Deposited metal (not cast iron) with requirements on machinability		
Cold arc welding, with beads: with electrodes based on copper and nickel	Small and medium blind defects on surfaces being machined and on surfaces after machining. In some cases short, through defects	Good machinability, density and strength (copper-nickel alloy)
with electrodes based on iron and nickel	Small and medium defects on various surfaces of castings and components	Good machinability, high strength and density parameters (iron-nickel alloy)
with electrodes based on low carbon steel with special coating	Small blind defects on machined surfaces	The hardness of the parent and the deposited metal must be identical (high-alloy steel)
with electrodes based on copper and steel	Through defects on unmachined surfaces of castings and walls of containers	High strength and density (copper-steel alloy)
With thin wires based on nickel or copper (multi-layer)	Through defects in thin walls of components	Good machinability, density and strength (nickel or copper alloy)
Gas flame low-temperature (without melting of parent metal) braze welding: with filler materials of copper alloys	Small blind defects on machined surfaces	Sufficient density and hardness 170-190HB (alloy of type LOMNA)
with self-fluxing powder alloys	Small and medium blind defects revealed during the machining of castings	The hardness of the parent metal and the deposited metal must be identical (nickel-boron-silicon alloy)
Metallising	Porosity on machined surfaces, sealing of leaking areas	(Alloys based on Mo, Zn, etc.)
Deposited metal (not cast iron) without requirements on machinability		
Cold manual arc welding with steel electrodes or semi-automatic welding with a steel wire	Blind defects on unmachined surfaces of castings	Good machinability (carbon steel at increased cooling rate)
Mechanical seals (plugs, inserts)**	Blind defects on machined surfaces separate - with plugs, groups of defects - with inserts)	(Cast iron)
Filling up	Blind defects, surface asperities, mainly on unmachined surfaces	(Hardening polymers)
Dipping (sealing)	Leaks in walls of castings	(Various substances)

* The prescribed properties of cast iron (ferritic-pearlitic, pearlitic, with spheroidal graphite) are obtained by using appropriate welding consumables

**These are used when welding cannot be employed

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machinability (with high nickel and copper contents, high-alloy steels). In addition, defects in cast iron components are repaired by mechanical seals (inserts, plugs, etc.), by filling up with hardening polymers, and by treatment with thickeners.

There are also a number of methods that can be employed in exceptional cases (resistance-powder, electroslag, strip surfacing, soldering with zinc alloy Ts at a temperature of 350°C; and also brazing with chemical copper plating at 200°).

The technological effectiveness of the method used plays an important role in the assessment of the suitability of application of specific methods of welding of cast iron taking into account stringent requirements on quality, reliability and economic effectiveness.

The characteristics of technological effictiveness are usually assessed by comparing the qualitative parameters of the welded joint (produced by a specific welding method) with technical requirements on the quality of cast iron components.

Welding is used most extensively in the repair of defects in iron castings before and after machining. In this case welding is regarded as an integral part of casting technology.

Necessary industrial bases have been constructed for welding; these are equipped with means of heating cast iron components of various dimensions, and with equipment for manual and mechanised processes. The required range of welding consumable is available. Work is carried out by qualified welders.

In plants with such bases the qualitative parameters of repaired castings and components meet the requirements of the technical conditions.

The fact that welding production is distributed over many plants that produce and repair cast iron castings and components, and in many cases do not have specialised departments, forces industrial workers to find compromise solutions between the need to recondition cast iron castings and components and the technical requirements imposed on the quality of reconditioned cast iron components. In these case the cold welding methods could play a leading role, because they enable production problems to be solved; but, taking into account the specific properties of cast iron, these methods do not make it possible to produce joints of the required quality. The fact is that in the arc welding and surfacing of cast iron without preheating, local thermal effects, occurring during the melting and solidification of cast iron, cause pronounced phase changes in zones adjacent to the welding zone. In this case the phase changes may be metastable, and this facilitates the formation of ledeburitic and martensitic structures, which are associated with the accelerated cooling of molten metal.

The formation of ledeburitic structures and the occurrence of martensitic transformations in the welding of cast iron can be prevented by the slow cooling of castings and components, heated in the first case to 600-650°C and to 150-400°C in the other case. These conditions are difficult to fulfil in the cold welding of cast iron because the conditions of heat exchange are unfavourable, and this results in the formation of hardened interlayers in the weld zone and in stress concentrations in the heat affected zone (HAZ). But these shortcomings are eliminated by the application of hot welding, which is already being used on a wide scale.

CONCLUSIONS

1. Welding with preheating (and, in some cases, with accompanying heating) occupies a leading position in the repair of iron castings in casting production.
2. In the reconditioning of worn components, the principal requirements usually concern the strength parameters. Welding without preheating can be applied in these cases.

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Welding cast irons

By E.N. Gregory, FIM, FWeldI, and S.B. Jones, BSc, PhD

An outline is given of the solidification and solid state transformations possible in cast irons, and the metallurgical factors that underlie a successful approach to cast iron welding are established. The choice of welding process and consumables is discussed.

INTRODUCTION

The term 'cast iron' covers a large range of iron/carbon alloys with widely differing structures and properties. These materials are of considerable industrial importance, annual UK production being second only to steel. Nowadays, cast irons are finding applications in competition with other materials such as steel castings or forgings. This has been made possible by improvements in alloy design, foundry techniques, and heat treatments, coupled with the inherent flexibility of the casting process. In parallel with this development there is a growing demand for cast iron welding, whether for repair or fabrication, in all branches of engineering. As most welding processes involve localised heating and fusion, the solidification and solid state transformation behaviour of cast irons provides the key to their weldability. An understanding of these factors is therefore essential in a rational approach to cast iron welding.

VARIETIES OF CAST IRON

Cast irons may be divided into two main classes: those which contain graphite, and white irons. Of these, the graphite-bearing irons are of the greatest commercial importance and include flake graphite, spheroidal graphite (ductile irons), and malleable irons.

Grey cast irons

Grey irons form a series containing 2.0-4.5%C and 1.0-3.0%Si. Their structures consist of branched and interconnected graphite flakes in a matrix which is commonly pearlite, ferrite, or a mixture of the two, Fig.1.

BS 1452 : 1961* contains seven grades of grey iron, classified by tensile strength. Typical properties of these grades are shown in Table 1a.

Spheroidal graphite (SG) iron

The mechanical properties of grey irons may be greatly improved if the graphite shape is modified to eliminate planes of weakness caused by continuous flakes. Such a modification is possible if molten iron having a composition in the range 3.2-4.5%C and 1.8-2.8%Si is treated with Mg or Ce additions and inoculated with Si before casting. This produces castings with graphite in spheroidal form instead of flakes, Fig.2, and these are known as nodular, SG, or ductile irons (BS 2789 : 1961**).

From their introduction in the late 1940s SG irons have become increasingly important, finding demanding applications in competition with steel forgings or other types of cast iron. SG irons are available with pearlite, ferrite, or pearlite/ferrite matrices (austenitic irons will not be considered), which offer a range of ductilities and tensile strengths, Table 1b, considerably better than those of grey irons. The characteristics of SG irons are given below.

Pearlitic SG iron

The pearlitic condition is the normal state of ordinary as-cast SG iron in sections of up to about 50mm. Much of the carbon is present

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* BS 1452 : 1961 'Grey iron castings'

** BS 2789 : 1961 'Iron castings with spheroidal or nodular graphite'.

Source: Welding of Castings, The Welding Institute, 1977

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Table 1 Typical mechanical properties of cast irons

Type	Grade	Min. tensile strength, N/mm^2	0.5% proof stress, N/mm^2	Elongation, %
(a) Grey iron	10	155	-	-
BS 1452 : 1961	12	186	-	-
	14	217	-	-
	17	263	-	-
	20	310	-	-
	23	356	-	-
	26	400	-	-
(b) Nodular iron	SNG 24/17	371	232	17
BS 2789 : 1961	27/12	418	279	12
	32/7	495	340	7
	37/2	572	387	2
	42/2	650	433	2
	47/2	730	464	2
(c) Whiteheart malleable iron	W410/4	410	250	4
BS 309 : 1972	340/3	340	-	3
(d) Blackheart malleable iron	B340/12	340	200	12
BS 310 : 1972	310/10	310	190	10
	290/6	290	170	6
(e) Pearlitic malleable iron	P690/2	690	540	2
BS 3333 : 1972	570/3	570	420	3
	540/5	540	340	5
	510/4	510	310	4
	440/7	440	270	7

in the combined form and the microstructure consists of intimately mixed ferrite and cementite with spheroids of graphite, giving rise to a hard, strong iron with about twice the strength of ordinary grey cast iron. The type corresponds to BS 2789, SNG 32/7, SNG 37/2.

The mechanical properties of as-cast SG iron are improved by normalising, that is, by first heating the castings to a supercritical temperature, holding for a sufficient time, and then allowing them to cool freely in air. Complex castings, and parts where maximum dimensional stability is required, also benefit from stress relief heat treatment at 550°C, whether or not this is preceded by the normalising treatment.

Ferritic SG iron

When pearlitic SG iron is annealed the matrix structure becomes ferritic; the combined carbon is broken down and is deposited on to the graphite spheroids in the ferritic matrix. In this condition the iron exhibits maximum toughness and ductility, with increased elongation. Material having a ferritic or predominantly ferritic matrix is covered by BS 2789, SNG 24/17 and 27/12. The matrix structure and production heat treatment of the above types of SG iron are shown in Table 2.

Malleable irons

All malleable irons are produced by heat treatment of white iron castings. The eutectic carbide of the white iron is decomposed to give carbon aggregates dispersed in a matrix which may be ferritic or pearlitic, Fig. 3. Malleable iron may be divided into two main classes: whiteheart and blackheart.

Whiteheart malleable iron

Whiteheart malleable castings are annealed in a decarburising medium. Carbon is removed at the casting surface, the loss being only partly compensated by diffusion of carbon from the interior. After this treatment, whiteheart castings are inhomogeneous with a decarburised surface skin and a higher carbon core. The degree of surface decarburisation depends on

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Table 2 Structures and production heat treatment of SG iron

BS grade	Matrix structure	Production heat treatment
24/17 27/12	Ferrite with traces of pearlite	Austenitised at 900°C - slow cool to 700°C - soak - furnace cool
32/7 37/2	Pearlite and ferrite	May be as-cast or heat treated depending on production conditions
47/2 42/2	Pearlite	Austenitised at 900°C followed by cooling in still air

section size: regions below 5mm thick may be almost totally carbon free.

Whiteheart malleable irons are covered by BS 309 : 1972.* Typical properties are shown in Table 1c.

Blackheart malleable iron

Blackheart irons (BS 310 : 1972**) are produced by annealing white iron castings without decarburisation. The resulting structure, which consists of carbon in a ferritic matrix, is homogeneous irrespective of section size. Mechanical properties are generally superior to whiteheart irons as shown in Table 1d. Blackheart irons are, however, more difficult to cast as the white iron used to produce them usually has a lower carbon content than whiteheart castings.

Pearlitic malleable iron

This type of malleable iron may be produced either by conventional blackheart annealing, but using a white iron which has been treated with carbide stabilisers, or by a further heat treatment of ferritic blackheart malleable iron. Pearlitic malleable castings (BS 3333 : 1972***) have increased strength but lower ductility than ferritic blackheart malleable irons, Table 1c.

WELDING AND CAST IRON METALLURGY

Although there are many varieties of cast iron with wide ranging compositions and mechanical properties, they share certain metallurgical features which must be controlled if they are to be welded successfully.

(a) Solidification

The formation of white iron eutectic during welding will markedly reduce component performance. Its formation must therefore be kept to a minimum.

(b) Solid state transformations

Brittle structures such as bainite and martensite in cast iron HAZs may cause underbead cracking if these regions are subjected to contraction stresses as the weld cools. The extent and hardness of these structures must therefore be limited.

(c) Weld metal dilution

Fusion welding will result in mixing of molten cast iron with the filler metal. Consumables must be chosen to be compatible with the dilution levels encountered with each welding process.

Control of fusion and solidification

In the foundry, cast iron solidification is controlled through alloy composition, inoculation to influence nucleation, and physical factors such as pouring temperature and choice of mould material. The scope for solidification control during welding is much more limited because of the rapid and localised nature of the heating and cooling cycles. Even with high preheats and correctly chosen filler alloy, white iron formation can seldom be completely avoided if welding causes fusion of graphite-bearing material. The structure of the white iron is usually acicular, indicating high cooling rates through the solidification range, Fig.4.

In view of this, cast iron arc welding techniques use low heat inputs, thus limiting the extent of parent metal fusion and consequent difficulty with weld metal dilution and brittle white iron zones. The aim of low heat inputs in fusion welding is to produce discontinuous white iron colonies in the fusion zone, thus lessening their weakening effect on the weldment, Fig.5. An alternative approach to the solidification problem is to avoid parent metal fusion altogether as in the bronze welding or powder welding techniques.

* BS 309 : 1972 'Whiteheart malleable iron castings'

** BS 310 : 1972 'Blackheart malleable iron castings'

*** BS 3333 : 1972 'Pearlitic malleable iron castings'

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Table 3 Typical preheat levels for welding cast irons

Material	Cast iron type	MMA, °C	MIG, °C	Gas fusion, °C	Powder welding, °C
Ferritic	Flake	300	300	600	300
	Nodular	RT-150	RT-150	600	200
	Whiteheart malleable	RT (200°C if high C core involved)	RT	600	200
	Blackheart malleable	RT-150	RT-150	600	200
Pearlitic	Flake	300-330	300-330	600	350
	Nodular	200-330	200-330	600	300
	Malleable	300-330	300-330	600	300

Control of HAZ properties

The formation of hard and brittle HAZ structures may lead to cracking during cooling or in service, and can cause problems if components have to be machined after welding, or if the welded joint has to comply with the ductility requirements of a standard specification or code which is determined by means of a transverse bend test, e.g. welds in SG iron pipes. Limitation of welding heat inputs reduces the extent of the HAZ formed, but may not always be enough to ensure satisfactory properties. HAZ hardness may be reduced by preheating followed by slow cooling after welding. Preheating lowers the cooling rate both in the weld and HAZ during and after austenite transformation, reducing martensite formation to a minimum and limiting the hardness of any which does form.

Cracking resulting from unequal expansion may occur during the preheating or complex castings, or when preheat is confined to small areas of large components. A general scheme of preheat levels is shown in Table 3.

Ferritic SG iron castings up to about 15mm thick can be welded without preheat. The general rules of preheating are not followed in the 'quench welding' technique for repairing iron castings. In this method the weld is made without preheat by depositing a series of small stringer beads. These weld beads are peened to relieve shrinkage stresses, and the weld area is quenched with an air blast or damp cloth with the object of limiting stress buildup and reducing HAZ size.

Recently at the University of Ghent aluminium alloyed/calcium inoculated cast iron has been developed in which silicon, the graphitising element of conventional grey irons, is replaced by aluminium in combination with calcium as an inoculant. This type of cast iron is claimed[1] to have good weldability without the formation of iron carbides in the HAZ.

Postweld heat treatment

Weldments containing white iron and martensite in their fusion and HAZs will have properties which are considerably inferior to similar components with homogeneous structures. The reduction in component properties caused by welding is unacceptable in many instances, even when components are designed with welding in mind. In such situations it may be possible to improve the properties of the welded joint by postweld heat treatment. White iron can be decomposed by heating for a minimum of 3hr at 900°-950°C, but it will be appreciated that high temperature heat treatments will add considerably to overall welding costs, and may be impractical owing to problems with component size or distortion. If annealing is not carried out HAZ properties may be improved by stress relief at 620°C immediately after welding, followed by furnace cooling to 350°C. This treatment decomposes martensite and gives a softer structure while reducing residual stresses in the component.

Buttering techniques for control of dilution and metallurgical quality

Buttering involves the preparation of a joint for welding by depositing one or more layers of weld metal on the joint edges, using a low heat input technique under closely controlled conditions. The buttered surface is more weldable than the raw casting, and the cast iron is shielded from the immediate thermal shock of the welding process used to complete the joint. Fusion and HAZ problems are therefore restricted to the buttering runs, which

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may be made with compatible but expensive filler materials followed where necessary by annealing to remove carbides. Less expensive filler and higher heat inputs may then be used to complete the weld, often without preheating.

Buttering can be applied to SG iron in the foundry before the production heat treatment. Since the subsequent final welding operations affect only the overlay deposits, no massive carbide is formed in the fusion zones and no postweld heat treatment is required to obtain maximum ductility. It is reported that satisfactory ductility is obtained[2] when a combination of stainless steel buttering and basic coated ferritic steel filler is used for welding SG iron.

CHOICE OF CONSUMABLES

Several factors must be considered when choosing filler materials for cast iron welding. The requirements for an ideal filler material may be mutually exclusive in any particular instance but the following factors must be considered:

1 Cost
2 Matching strength
3 Tolerance to dilution by parent metal
4 Machinability
5 Tolerance to high cooling rates
6 Weldability at low heat inputs
7 Colour matching
8 Ductility to absorb welding strains

Common filler metals for welding cast iron include SG iron, nickel-base alloys, and copper-base alloys. These are described in detail in the next Section.

CHOICE OF WELDING PROCESS

Oxyacetylene welding

Oxyacetylene welding produces very low penetration and dilution, and, to melt the surface of the cast iron with an oxyacetylene gas flame, it is general practice to preheat the iron to 600°-650°C. Although this high heat input results in a wide HAZ, the very slow cooling rate after welding causes the HAZ to transform to soft microstructures without any risk of martensite formation.

Filler materials generally available for oxyacetylene welding include SG cast iron rods[3] having high contents of carbon and silicon which promote graphitisation of the weld metal as it solidifies and cools, with consequent good machinability. For higher duty castings, higher strength filler rods are available containing alloying elements such as nickel and molybdenum.

Despite the above advantages of oxyacetylene welding, it is a very slow process and therefore largely unsuitable for production welding. Its use is restricted to the repair of complex castings.

Bronze welding

Bronze welding is a nonfusion process in which the parent metal is not melted and a brass filler rod is used. This process is referred to as braze welding in the USA. The parent metal is preheated either locally or throughout the casting to 400°C and the region of the joint is coated with flux. The part at which welding is to commence is then preheated with the oxyacetylene flame to a temperature of approximately 850°C. The filler rod is warmed with the flame, dipped in the flux, and then applied to the heated casting when it should immediately melt and wet the surface of the casting. Further filler metal is added as welding proceeds until the prepared groove is filled. Bronze welding gives a machinable joint that is suitable for service up to 250°C, at which temperature the strength starts to fall.

There is no colour match but this is often not important and the reduced heat input compared with full fusion welding with the oxyacetylene process may in certain situations be useful in avoiding distortion. As with correctly applied oxyacetylene fusion welding, the bronze welding process produces a soft HAZ free from carbides and martensite.

Powder welding

Powder welding is a low heat input gas welding process somewhat akin to bronze welding in respect of parent metal temperatures but utilising a number of proprietary filler metals in powder form contained in a closed hopper mounted on the specially designed oxyacetylene welding torch. The finely divided powder is fed through a metering valve into the oxygen stream so that it passes through the gas flame before it impinges on the surface of the workpiece where surface alloying occurs. The presence of silicon and boron in the powder alloys lowers the melting point and produces a self-fluxing action which facilitates wetting of the cast iron parent metal. As with bronze welding, the parent metal is not melted by the oxyacetylene flame and therefore carbide formation is suppressed. Provided that the cooling rate after welding is controlled a soft HAZ will be obtained.

The advantage of powder welding over bronze welding is that higher strength weld metals are available whose use is not restricted to service temperatures below 250°C. Many

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different powder alloys are available, most of them being based on nickel with additions of iron, chromium, and cobalt which produce weld deposits having a range of hardnesses and structures giving different degrees of wear resistance. These filler metals are suitable for a variety of applications and can be used to give high strength welds in grey, SG, or malleable irons; some of them can be used for wear-resistant weld deposits.

A useful technique is to use powder welding for buttering the edges of a joint, so that a soft HAZ is obtained, followed by the use of a faster process such as manual metal-arc (MMA) welding to complete the joint.

MMA welding

Manual metal-arc welding with covered electrodes has been applied to the repair of cast iron for many years and is also used for welded fabrications incorporating cast iron.

The development of suitable electrodes has been influenced by the metallurgical characteristics of cast iron and also by the particular features of the welding process. Whereas with gas welding processes some degree of preheat is essential to raise the temperature of the joint faces so that the oxyacetylene flame can melt the surface more easily, the act of striking an arc on cast iron causes a small molten pool to form immediately, whether preheat is used or not, and however thick the casting.

The penetration or amount of parent metal melted is much greater with MMA welding than with the gas welding process and this molten cast iron becomes mixed with the filler metal deposited from the electrode. Depending on the type of electrode and the welding current the amount of dilution by parent metal can be up to 30% and this has influenced the development of electrodes. For example, if a normal mild steel electrode is used on cast iron the weld deposit will contain about 1% carbon and, at the comparatively rapid cooling rates of arc welds, the deposit will transform to martensite which, because of its inherent brittleness, is almost certain to crack under the action of contraction stresses. This quench cracking is likely to occur even when using a basic covered electrode which has been properly dried at temperatures of 450°-500°C before welding.

Nickel electrodes

The most satisfactory solution to the problem of weld metal dilution by carbon from the parent metal is the use of a nickel or nickel alloy consumable. The carbon separates in the form of finely divided graphite and a readily machinable weld deposit is obtained. The weld metal is ductile even when welding without preheat because no transformation takes place such as occurs in a ferritic deposit. Rapid cooling rates, therefore, do not have a deleterious effect on the weld metal. One disadvantage of nickel weld metal is its susceptibility to solidification cracking from pickup of phosphorus or sulphur from a cast iron containing high percentages of these elements. Spheroidal graphite irons are normally low in these elements and this problem does not occur.

Nickel electrodes for welding cast iron classified as ENi-CI by AWS A5.15.69 have a graphite-based covering which produces a soft shallow penetrating arc enabling the electrode to be used for low current welding with either AC or DC. Although the nickel weld deposit is not affected, preheat can influence HAZ properties, as discussed in 'Control of HAZ properties'. Such electrodes may be economically used for the deposition of buttering layers.

Nickel/iron electrodes

These electrodes, classified as ENiFe-CI by AWS A5.15.69, contain approximately 55% nickel and 45% iron. They are cheaper than pure nickel electrodes and are less susceptible to solidification cracking caused by phosphorus or sulphur. The weld metal hardness is higher than that of pure nickel and because it is increased by dilution this must be controlled to obtain reasonable machinability of the weld. Preheating to 150°-300°C will reduce the cooling rate of the weld and will improve machinability.

These electrodes produce stronger weld metal than nickel electrodes and are widely used for high strength welds in SG irons and high duty irons and also for welding cast iron to steel. They can also be used for buttering joint faces as described for nickel electrodes.

The mechanical properties of butt welds made in pearlitic and ferritic SG irons with nickel-iron electrodes are compiled in Table 4 published by the British SG Iron Producers Association Ltd.[4] These results are based on the work of several investigators and are typical of the mechanical properties which can be expected when joining SG iron castings in the pearlitic and ferritic grades to one another. They indicate that reasonably strong and moderately tough and ductile welds can be obtained by metal-arc welding.

From the Table it is also evident that ductility is improved by postweld annealing heat treatment which eliminates carbides and

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Table 4

Grade of iron	Condition of cast plate	Mechanical properties							
		Yield strength, N/mm^2	Maximum tensile strength, N/mm^2	Elongation, %	Unnotched Charpy impact strength, J*	Hardness, HV		Bend angle, ° ≠	Nil ductility transition temperature from drop weight specimens, °C
						Weld metal	Maximum in HAZ		
Ferritic	Unwelded	230-307	369-461	12-20	130-138	-	-	78-80	-85
	As-welded	277-400	354-477	1-9	18-26	205-250	650-700	14-28	-
	Stress-relieved, 600°-650°C various times	246-400	369-492	5-14	36-47	185-235	400-430	18-38	-
	Full ferritise anneal, 2-4hr 900°C, 8-16hr 700°C, furnace cool	277-354	415-461	6-12	39-47	175-185	160-180	40-45	-71
Pearlitic	Unwelded	461-600	569-723	3-5	13-16	-	-	5-10	-
	As-welded	461-523	461-569	1-3	16-20	190-225	650-700	14-16	-
	Stress-relieved 600°-650°C various times	461-492	431-508	1-3	13-16	175-205	420-500	-	-
	Full ferritise anneal, 2-4hr 900°C, 8-16hr 700°C, furnace cool	338-369	400-446	6-12	47-52	175-180	180-190	36-42	-

* Tests made at room temperature. ≠ The bend angle quoted covers specimens of different shapes and with both weld face and weld root used as the point of application of load. In this test a specimen 356mm long is supported on rests and a weight of standard size is dropped from a specific height. A notch is provided usually by placing a brittle weld bead across the centre of the specimen which introduces a running crack after about 3° bend. A stop prevents bending beyond 5°. Thus, if the specimen breaks during the test the mechanism of failure is primarily cleavage. Complete failure of drop weight specimens occurs at temperatures below the 'nil-ductility transition temperature'.

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Table 5 Transverse tensile properties of butt welds made by MIG-welding 25mm thick pearlitic and ferritic SG iron plates

Plate	Filler metal	Transverse tensile properties*			Location of fracture
		0.2% proof stress, N/mm^2	Maximum stress, N/mm^2	Elongation over weld, % 50mm	
Pearlitic as-cast	Unwelded	384–461	615–769	1–3	-
	Nickel 61	357	520	3.0	FL and HAZ ≠
		357	575	4.0	-
Fully ferritic	Unwelded	230–308	385–538	15–25	-
	Nickel 61	302	425	11.5	Away from weld
		295	415	11.5	Away from weld

* Testpieces 14mm diameter over 80mm parallel portion
≠ FL - weld fusion zone, HAZ - heat-affected zone

martensite from the HAZ. Any improvement in ductility tends, however, to be limited by the distribution of the graphite found from the breakdown of carbides and martensite.

Nickel/copper electrodes

These electrodes of the ENiCu-A and ENiCu-B types are often referred to under the trade name of Monel.* The covering is similar to that of nickel or a nickel/iron electrode and is based on graphite to give low penetration and hence decreased dilution. This is particularly important for a Monel filler metal because nickel/copper alloys are susceptible to cracking when diluted by iron.

Copper-base electrodes

These electrodes classified under AWS A5.15.69 as ECuSn and ECuAl-A2 consist of tin bronze (copper/tin) or aluminium bronze (copper/aluminium) alloys. They were developed many years ago for welding tin bronzes, aluminium bronzes, or brasses and it was subsequently found that they could be used for welding cast iron without the problems associated with cast iron or steel electrodes. These electrodes have been largely superseded by the nickel and nickel/iron types.

MIG-welding

In recent years MIG-welding with nonferrous filler wires and an argon gas shield has been applied to the welding of cast irons. The low heat input dip transfer or short circuiting arc process generally produces less iron carbide in the HAZ than MMA welding. Another advantage of MIG-welding is that it can be mechanised and is therefore ideal for repetition work on welding cast iron components. Once the optimum welding conditions have been determined work should proceed without trouble, provided the composition of the cast iron does not vary to an unacceptable degree.

There are no specifications for filler wires for MIG-welding cast iron but satisfactory results have been obtained with Nickel 61 and Monel 60 wires (developed by the International Nickel Company) which are covered by BS 2901 : Pt 5 : 1970 and AWS A5.14.69. The compositions of these wires are given in Table 5. Nickel/iron

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Table 6 Composition of nickel and nickel/copper filler wires for MIG-welding; covered by BS 2901 : Pt 5 : 1970, NA 32 and NA 33 and AWS A5.14.69, ERNi-3 and ERNiCu-7

Trade name	Chemical composition, %								
	Nickel	Iron	Titanium	Aluminium	Manganese	Carbon	Silicon	Copper	Sulphur
Nickel 61	93min.	1.0max.	2-3.50	1.50max.	1.00max.	0.15max.	0.75max.	0.25max.	0.01max.
Monel 60	62-69	2.5max.	1.5-3.00	1.25max.	1.00max.	0.15max.	1.25max.	Remainder	0.015max.

Table 7 Composition of copper alloy filler wires for MIG-welding

Specification	Chemical composition, %			
	Copper	Aluminium	Tin	Manganese
BS 2901 : Pt 3 : 1970, C10 AWS A5.6.69, ECuSn-A	93.8min.	0.03max.	4.5 6.0	-
BS 2901 : Pt 3 : 1970, C11 AWS A5.6.69, ECuSn-C	92.3min.	0.03max.	6.0 7.5	-
BS 2901 : Pt 3 : 1970, C13 AWS A5.6.69, ECuAl-A2	86.0min.	9.0 11.0	-	1.0max.

wires similar in composition to ENiFe-CI in Table 1 have been used successfully for MIG-welding but these are not readily available.

The application of MIG-welding to SG iron with nickel-base fillers and argon shielding was first described by Osman and Stephenson who found that the low heat input produced a discontinuous network of carbide which was restricted to discrete areas surrounding the graphite nodules. Good tensile properties matching those of the parent metal could be achieved without annealing when Nickel 61 filler wire was used, Table 6. The data given in Table 6 show that elongations of 11.5% can be achieved in as-welded ferritic SG iron.

Copper-based filler metals such as the tin and aluminium bronzes, Table 7, and also the manganese aluminium bronze Superston 40 (BS 2901 : Pt 3 : 1970, C22) (developed by Stone Manganese Marine Ltd) can be used for MIG-welding cast iron. The nickel- and copper-based filler wires can be deposited by either the dip transfer process or by the more recently developed pulsed arc method. Both these welding processes restrict dilution to the region of 10% and the low controlled heat input produces a minimum of brittle iron carbides.

In certain circumstances a ferritic filler wire can be used; for example, when welding a whiteheart malleable iron in which the low penetration of the dip transfer or pulsed arc process does not penetrate the ferritic skin of the casting.

For dip transfer welding with a ferritic filler wire, CO_2 shielding gas could be used in place of argon and thus offer an economic advantage. Procedural tests would be necessary to determine the preheat required which could vary from zero to 200°C depending on whether

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the HAZ included the high carbon core of the casting.

A flux-cored electrode for gas-shielded welding has been developed recently which deposits weld metal having a composition similar to that of the ENiFe-CI covered electrode. This new flux-cored electrode, NI-ROD* FC55, was designed for the high speed production welding of cast irons and it is reported[5] that it can be used without shielding gas, with CO_2 or CO_2/argon shielding, or with a submerged-arc flux.

It is likely that, because of the precise control of the heat input possible with MIG-welding, particularly when mechanised, the process will be applied more extensively in the future for the production welding of cast irons.

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* NI-ROD is a registered trade mark of the International Nickel Company Inc.

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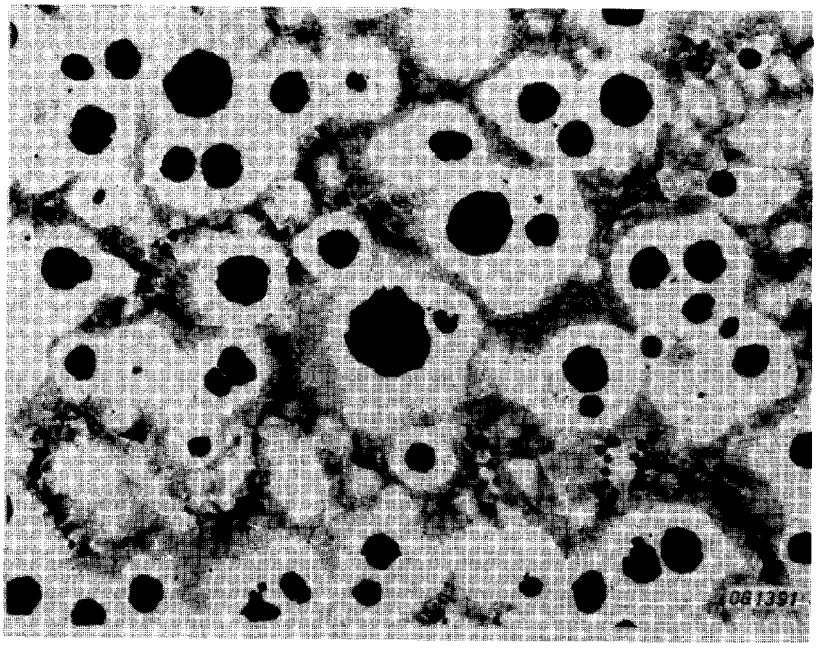
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1 Pearlitic grey iron x200

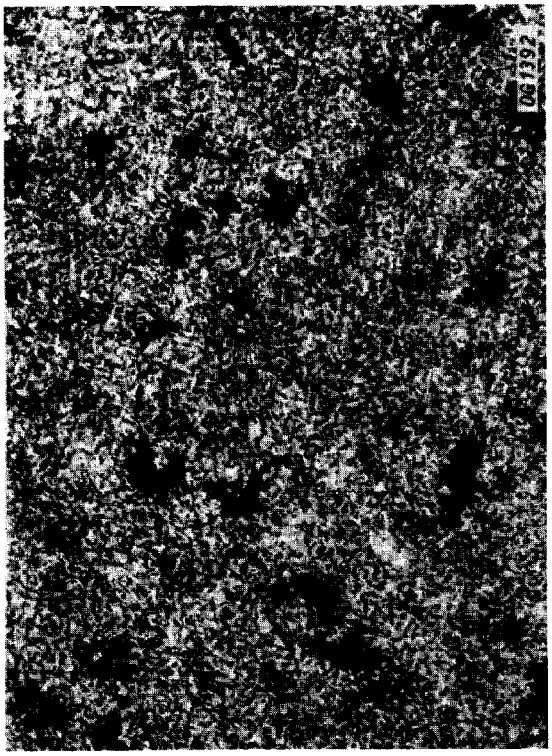


2 Nodular iron x200

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5 Discontinuous white iron colonies in fusion zone of blackheart malleable iron x 300



3 Pearlitic malleable iron x 130



4 White iron fusion zone structure x 1000

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The Development of a SMAW Procedure for Cast Iron

Implant test results lead to recommendation of E-NiCrFe-2 electrode and 700°F preheat and postheat

BY B. APPELT AND R. E. LONG

Introduction

This project entailed the development of a procedure for SMAW under field conditions of cast iron. The quantitative results provided an insight to SMAW procedures for cast iron. Variables of electrodes, preheats, and postheats were observed. The procedure was evaluated by the use of a test which was a modification of the implant test (Granjon, 1979) whereby threaded cast iron was welded to carbon steel plate. Each implant was tested by placing a nut on the threaded cast iron and using a torque wrench to determine the energy required to break the implant from the weld.

The problem was to determine what procedure would produce welds in cast iron with the highest strength when welded with the SMAW process under field conditions. This procedure was developed in a welding laboratory, but used only equipment applicable to field welding conditions. No furnaces were to be used with this process. Field conditions were to be a welding truck equipped with an engine-driven welding unit and oxyacetylene equipment.

Objectives of the Project

This study, dealing with the development of an optimum SMAW procedure, had the following objectives:

1. Determine a standard for cast iron when tested in this manner.
2. Determine suitable filler metals.
3. Determine suitable preheats and postheats.

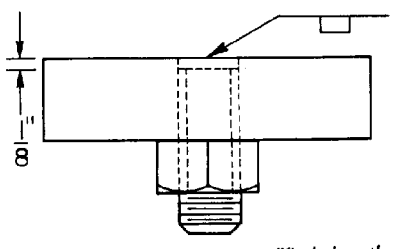


Fig. 1—Implant test as modified by the authors

An implant test was used as a means to gather information indicative of an acceptable welding procedure for cast iron by the SMAW process under field conditions. This test was originally developed by Granjon of France, and later modified by Sawhill and others of the USA (Savage, 1976). A further modification was made by the authors whereby torque, rather than direct tension was used for pulling the implant from the specimens. For this study a threaded piece of cast iron was welded to a low carbon steel base metal—Fig. 1.

Variables consisting of filler metals, preheats, and postheats were studied. The electrodes used in this study were E-NiCrFe-2 and E-Ni-1. The chemistry for these electrodes is stated in AWS A5.14 and is given in Table 1. The preheats used in this investigation were 500°F (260°C), 600°F (315°C), and 700°F (371°C). The postweld heat treatments (PWHT) used were also 500°F (260°C), 600°F (315°C), and 700°F (371°C).

All welding was done with an arc voltage of 18-20 volts (V) and 90-100 amperes (A) direct current reverse polarity using ⅛ in. (3.2 mm) diameter electrodes. After welding the samples were air-cooled and tested as welded or PWHT and tested. The evaluation test was accomplished by placing a nut on the threaded cast iron and using a torque wrench on the nut to determine the amount of energy required to remove each implant from its weld—Fig. 2.

Equipment and Materials

Equipment and materials used included:

1. Implants of 0.750 in. (19 mm) diameter cast iron threaded National Coarse.
2. Low carbon steel plates.
3. E-NiCrFe-2 and E-Ni-1 electrodes.
4. A heat source capable of preheating and PWHT of assemblies as required in the range of 500°F (260°C) to 700°F (371°C).
5. Nuts of 0.750 in. (19 mm) National Coarse.
6. A torque wrench.
7. A welding power supply with amp and volt meters.
8. Temperature indicating crayons to measure preheats and postheats.

Evaluation

Each weld was evaluated using the foot-pounds (Joules) of energy that each individual implant absorbed before cracking free of the plate. Comparisons were made between the welds to determine the most suitable filler metal, preheat and postheat for a field weld of cast iron.

No attempt was made to analyze the type of failure or preliminary microcracking. The evaluation was the complete energy level required to remove the implant of cast iron from the weld. The strength of the cast iron was determined by machining a special implant specimen (Fig. 3) with a flange on one end similar to a bolt head. This specimen was tested and found to require 170 ft-lb (230.5 Joules) of energy to break—Fig. 4.

Table 1—Chemical Compositions of Electrodes Used in Study, %

	E-NiCrFe-2	E-Ni-1
C	0.1	0.1
Mn	1.0-3.5	0.75
Fe	12	0.75
P	0.03	0.03
S	0.02	0.02
Si	0.75	1.25
Cu	0.5	0.25
Ni	62 min.	92 min.
Cr	13-17	–
Cb	0.5-3.0	–
Mo	0.5-3.0	–.
Al	–	1.0

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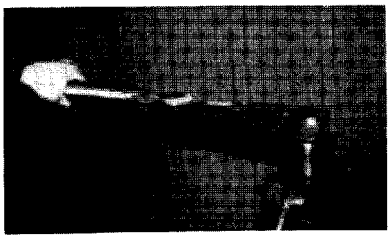


Fig. 2—Determining the energy required to remove implant from its weld

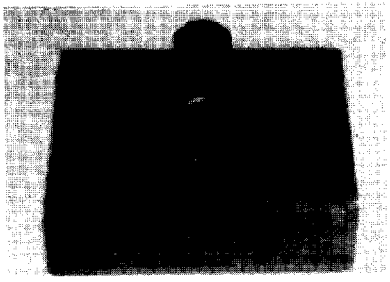


Fig. 3—Implant test specimen



Fig. 4—Fractured implant test specimen

Table 2—Implant Test Results

Temperature of Specimen when welded	E-NiCrFe-2	E-Ni-1
75°F (24°C)	30 ft-lb (40.68 joules)	48 ft-lb (65 joules)
Preheat only to 500°F (260°C)	38 ft-lb (51.5 joules)	57 ft-lb (77.3 joules)
Preheat and PWHT to 500°F (260°C)	102 ft-lb (138.3 joules)	63 ft-lb (85.4 joules)
Preheat only to 600°F (315°C)	83 ft-lb (112.5 joules)	60 ft-lb (81.4 joules)
Preheat and PWHT to 600°F (315°C)	122 ft-lb (165.4 joules)	98 ft-lb (132.9 joules)
Preheat only to 700°F (371°C)	92 ft-lb (124.7 joules)	57 ft-lb (77.3 joules)
Preheat and PWHT to 700°F (371°C)	138 ft-lb (187 joules)	92 ft-lb (124.7 joules)

Strength of unwelded implant was 170 ft-lb (230.5 joules)

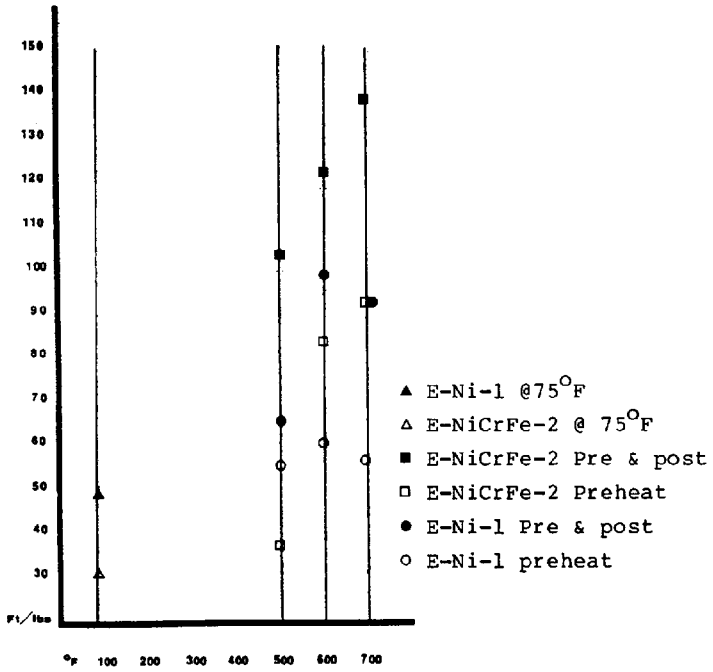


Fig. 5—The effects of various preheats and PWHT's on two electrodes

Results

The variables and results are presented in tabular form for simplicity, and all numbers are stated in foot-pounds and Joules. The strengths of all the various welds are presented in Table 2. The data are also presented graphically in Fig. 5 where the superiority of E-NiCrFe-2 with a 700°F (371°C) pre- and postheat may be seen.

The effect of pre and postheat on E-NiCrFe-2 is more easily seen in Fig. 6. The effect is an increase from 30 ft-lb (40.68 Joules) to 138 ft-lb (187 Joules). This is an increase of 460% from welding at 75°F (24°C). A weld made with E-NiCrFe-2 and using a 700°F (371°C) pre and postheat develops 81.17% of the base metal. The correlation coefficient for these tests was 0.99973, indicating strong correlation.

The use of preheat only produced the results illustrated in Fig. 7. This shows that

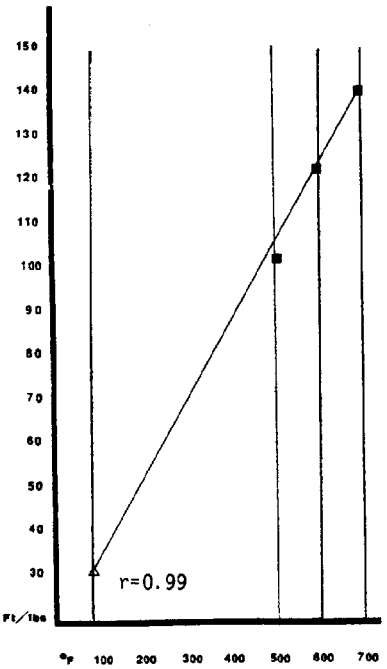


Fig. 6—Effect of increasing preheat and PWHT on E-NiCrFe-2

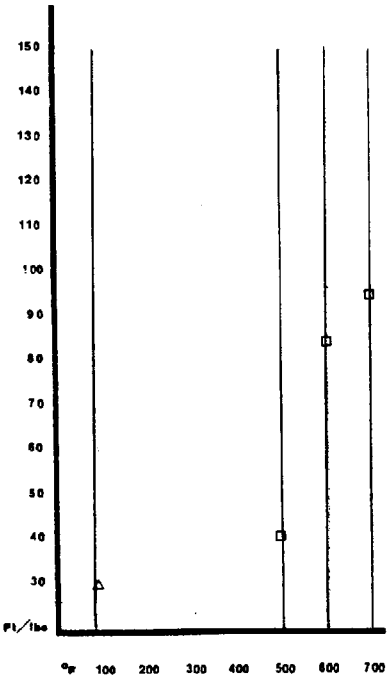


Fig. 7—Effect of increasing preheat on E-NiCrFe-2

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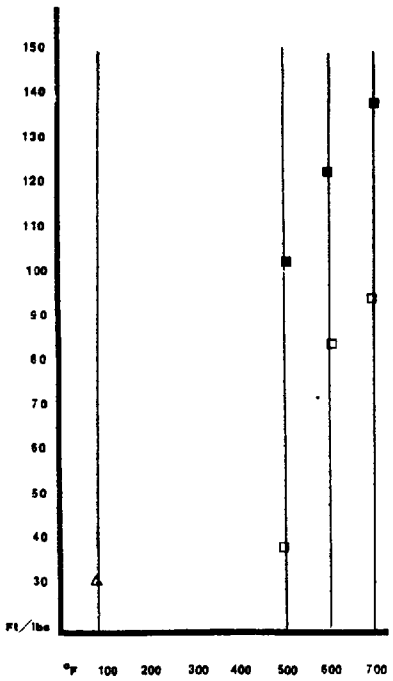


Fig. 8—Effect of preheat vs. preheat and PWHT on E-NiCrFe-2

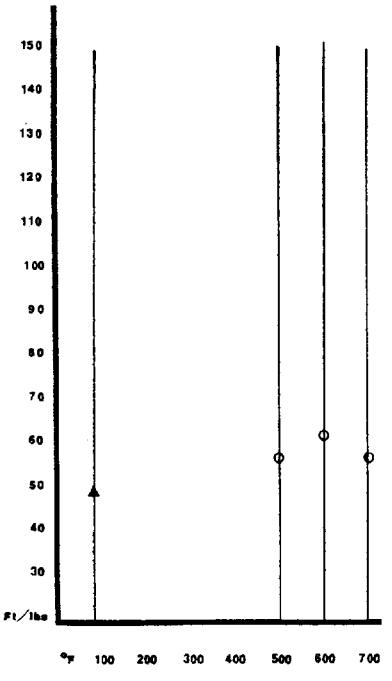


Fig. 9—Effect of increasing preheat on E-Ni-1

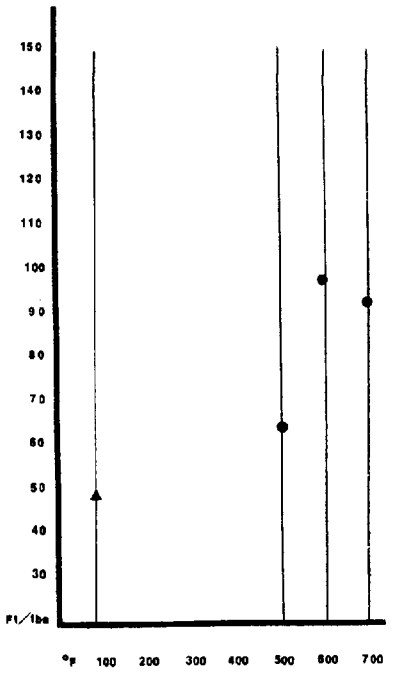


Fig. 10—Effect of increasing preheat and PWHT on E-Ni-1

there was little advantage to a 500°F (260°C) preheat alone, but 700°F (371°C) preheat did produce a strength of 92 ft-lb (124.7 Joules).

The results of using preheat and postheat with E-NiCrFe-2 are presented in Fig. 8. The 500°F (260°C) postheat caused a charge from 38 ft-lb (51.5 Joules) to 102 ft-lb (138.3 Joules). This is a charge of approximately 268%, but only 60% of the base metal strength.

The effect of preheat only on E-Ni-1 is shown in Fig. 9. Note that the strength of a weld with this filler metal is less at 700°F (371°C) than when welded with a 600°F (315°C) preheat.

The effect of preheat and postheat on E-Ni-1 is shown in Fig. 10. The weld made with 700°F (371°C) pre and postheat has a lower strength than that made with 600°F (315°C) pre and postheat. This is similar to the situation for preheat only.

Conclusion

The data indicate that the strongest weld on the basis of torque required for fracture was made with preheat and postheat using E-NiCrFe-2. An interesting point is that with increasing temperature there is an increase in torque strength of the specimens made with the E-NiCrFe-2. Further tests are needed to determine the ultimate strength of welds made with this electrode. There is an indication that the welds made with E-Ni-1 develop their ultimate strength when postweld heat treated to 600°F (315°C).

Based upon the data collected during this investigation, the welding procedure for cast iron under field conditions developing the greatest torque value on an implant test has three primary variables:

1. The electrode is E-NiCrFe-2.
2. The preheat is 700°F (371°C).
3. The weld is postheated to 700°F (371°C).

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APPLICATION TO CAST-IRONS OF VERTICAL WELDING UNDER SLAG (*)

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I. INTRODUCTION

There are currently two methods used in the few cases of the application of welding to the repair and construction of cast-iron parts:

- The so-called "hot thermal condition" or homogeneous welding method.

This technique consists of pre-heating the part to a high temperature (500°C approx.) which on the one hand gives the material sufficient plasticity to free it from thermal stresses and which on the other hand allows the cooling speed to be reduced, in order to produce a structure free of brittle metastable components.

- The so-called "cold thermal condition" or heterogeneous welding method.

By contrast with the previous method, this technique consists of using a small amount of energy to apply multiple deposits in order to limit the area affected by the thermal cycle to the minimum strictly necessary, the appearance of brittle metastable components being inevitable within this area.

In view of the lack of reliable and easily applied processes, we undertook the study whose results are given below, within the framework of a contract from the M.I.R.

(*) Paper presented to the Society of Welding Engineers in Paris on 15th November 1979.

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Vertical welding under slag offers certain peculiarities which should be mentioned at the outset. These are as follows:

- Very high energy input: the low speed of movement of the heat source produces a flattened thermal gradient with considerable natural pre-heating by conduction. This characteristic naturally classifies the process among the hot thermal condition welding methods.
- Creation of a considerably diluted massive molten area, insensitive to hot fissuring, because of the special orientation of the solidification.
- Favourable shape factor; absence of geometrical notching at the joint, associated with an area of extensive and homogeneous residual stresses.
- Symmetrical energy distribution around the centre of the weld and final solidification at the core.

II. DIRECTION OF THE STUDY

2.1 Objective

The main objective of the study was to obtain in the as welded state, welded joints suitably free from porosity and capable of being machined by machine-tools.

2.2 Grade and characteristics of the test plates.

Two non-alloy haematite cast-irons for fabrication were chosen:

. Lamellar graphite cast-iron F_T 25 D (NF A 32-101).
. Spheriod graphite cast-iron F GS 50.7 (NF A 32-201).

The chemical analysis of each grade is given in table I.

For each grade, test plates were supplied to thicknesses of 12, 20 40 mm, in a 600 x 250 mm rectangular format, which allowed:

- either the preparation of short samples, for the preliminary tests, technological development and study of structures;
- or the preparation of long samples, allowing the creation of a significant thermal gradient.

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2.3 Techniques studied

Two separate techniques were studied:

2.3.1 Welding with cast-iron rectangular filler metal rods

This technique consists of moving at a suitable speed a prismatic electrode rod of sufficient length and of section suitable for the thickness of the parts, in order to obtain a molten area of the same composition.

2.3.2 Welding with fusible wire-guide

This technique, which offers great simplicity of application, allows the use of various combinations of wires and wire-guides to be studied:

- solid wire of steel and wire-guide of cast-iron;
- coated wire depositing a cast-iron and wire-guide of cast-iron or steel.

In this manner, and as a function of the relative section of the wire-guide, the composition of the molten area can be adjusted to the required value.

TABLE I

	Product analysis					
	C total %	Mn%	Si%	S%	P%	Mg%
F_T25D	3.25	0.85	2.85	0.016	0.040	-
F GS 50.7	3.15	0.35	2.80	0.012	0.035	0.04

TABLE II

Ref.	CaF_2	CaO	MgO	MnO	FeO	Al_2O_2	TiO_2	SiO_2	Basicity
A	20	14	14	6	-	17	-	28	Neutral, weakly basic
B	45	12	2	2	2	20	5	6	Basic
C	16	19	10	6	-	15	-	33	Neutral, weakly acid

2.4 Equipment used

The various tests were performed with a view to using conventional equipment and consumables; thus:

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- the welding current generator was a transformer-rectifier with flat voltage characteristic capable of supplying a current of 600 A in continuous service;
- the electrode wires supplied were of conventional grades;
- the electrode rods and fusible wire-guides were machined from the 12 mm test plates.

2.5 Slags studied

The research and development of a slag offering certain physical characteristics suitable for the welding of cast-irons was outside the scope of the study and was therefore not undertaken.

Consequently, three slags of different compositions and used for the welding of steels, were chosen.

The mean chemical weight analysis of each slag is given in table II.

III. WELDING WITH RECTANGULAR ROD

3.1 Preliminary technical problems

Among the problems inherent in the use of this technique is that of the guidance and movement of the prismatic rod.

Figure 1 illustrates the rig studied and prepared by us. This rig, adaptable to the basic motor-reeler, permits the vertical movement of a square section rod of 12 mm sides at a speed adjustable from 0 to 35 cm/min.

Moreover, spark butt-welding was found particularly suitable for the preparation of rods in lengths of 1200, 1800 and 2400 mm.

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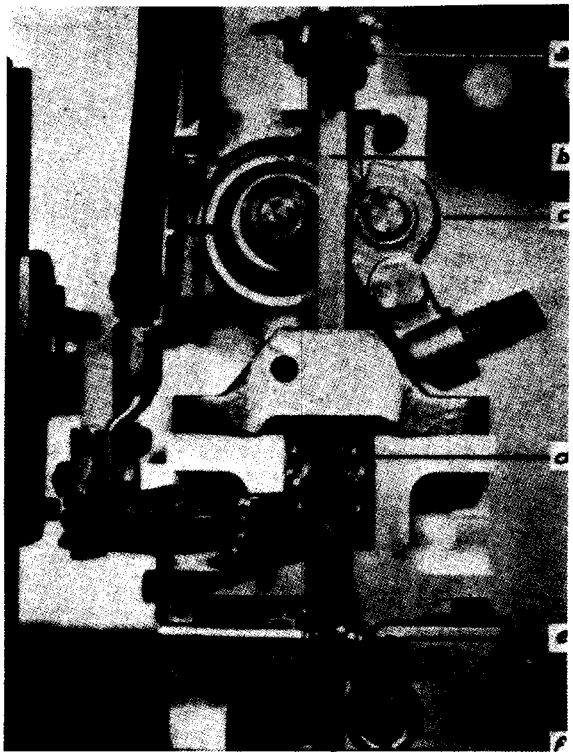


Fig. 1

a: top transverse guide roller
b: prismatic rod (12 x 12)
c: top side guide and driving rollers
d: electrical contact pad
e: bottom transverse guide roller
f: bottom side guide rollers

3.2 Welding test with mild steel rectangular rod

After several encouraging tests performed on test-pieces of 20 mm gauge E 24.1 steel, during which the three slags (A,B and C) were successively used, a transitional test was performed in order to study the behaviour of the liquid slag ref. A in the presence of cast-iron F_T 25.

From the porosity and weld metal inclusion standpoint, the result obtained was satisfactory.

However, the molten area offers a structure of hyper-eutectoid composition whose excessive hardness (380 to 400 Hv) prevents its machining by machine-tools.

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3.3 Welding tests with cast-iron rectangular rod

3.3.1 Preliminary tests

The choice of filler metal grade fell upon cast-iron F GS 50 whose low Mn content offers advantages from the graphitisation standpoint.

The first experiments conducted with slag A and cooled copper pads were disappointing from the point of view of the soundness of the joints. The visual examination in fact showed the presence of quasi-continuous slag inclusions whose scarfing started shortly after the fusion of the cast-iron rod.

The cross-sectional macrographic examination revealed a heterogeneity of the structure of the molten area (figures 2a and 2b).

The graphitic regions, of low hardness (190 to 230 Hv) revealed a fine interdendritic superfusion lamellar graphite of type D (micrograph in figure 3).

By contrast, the hardness of the carburized areas was excessive. (450 to 560 Hv).

Finally, the presence of spheriodal metal particles could be observed in the slag.

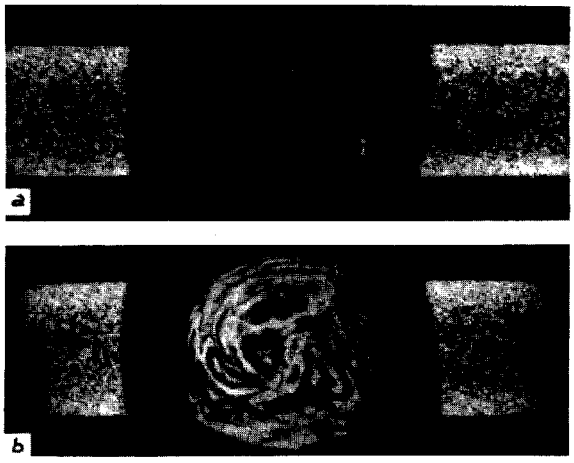


Fig. 2

F_T 25. g = 20 mm. Slag A.
Rod FGS50

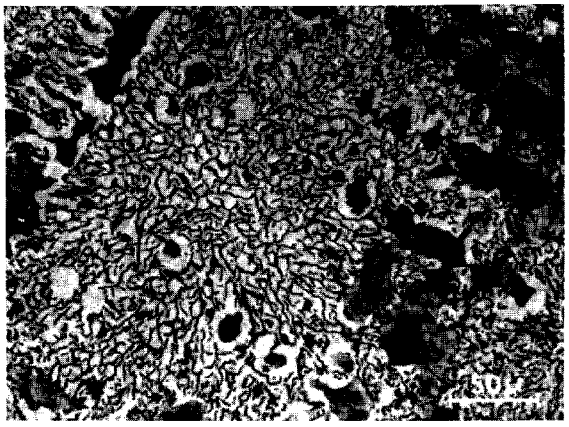


Fig. 3

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3.3.2 Complementary tests

The latent fault in weld metal soundness observed during the preliminary tests led us to study the action of certain parameters upon the behaviour of slag A:

. Increase of the welding energy:
  - use of refractory graphite pads;
  - increase in cross-section and alumina pads.

. Limitation of the quantity of slag.

. Slag enrichment with calcium fluoride.

Other parameters were also studied:

. Behaviour of the type B and C slags.

. Influence of the F_T 25 filler material on slag A.

None of the actions taken provided any remedy against the formation of the inclusions which remained just as dense.

The increased energy resulted in local fusion of the edges and the creation of a sizeable molten area more or less separated from the latter by inclusions of slag.

With slag C, the weld even gave the appearance of an actual ingot isolated from the edges by a thin layer of slag.

The use of F_T 25 lamellar graphite cast-iron as the filler material was additionally characterised by considerable instability of the electrical parameters.

3.3.3 Final test with composite rod

The behaviour differences found between steel and cast-iron electrode rods led us to suspect the graphitic carbon. Consequently, a decrease in the graphite concentration of the electrode rod should make it possible to lower the reactivity of the carbon with the oxides of the slag.

The test performed along these lines made use of a rod consisting of an F GS 50 cast-iron core of rectangular section (9 x 12 mm), plated on two sides with a 1.5 mm thick ribbon of extra-mild steel.

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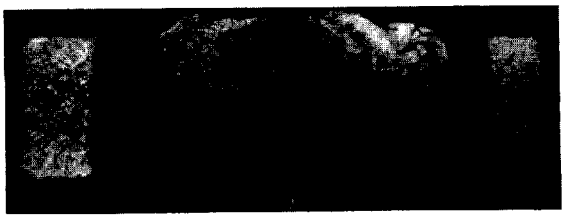


Fig. 4.
F_T25. e = 20 mm. Laitier A. Barre composite FGS50 plaquée acier doux.

Fig. 4

F_T25. g = 20 mm. Slag A. Composite rod FGS50 plated with mild steel.

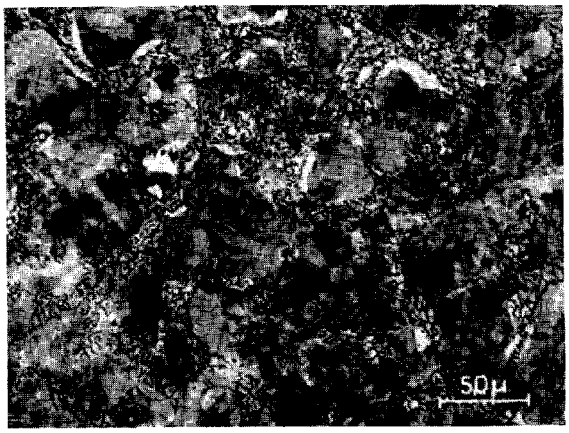


Fig. 5 (opposite).

The result obtained was finally crowned with success.

Despite the modest welding energy value, no weld defects were able to be detected between the molten area and the base metal.

The macrographic examination showed a mixed structure, graphitic at the heart and cementic at the edges of the cooled pads (fi.g 4).

The micrographic examination of the cove again revealed a fine interdendritic lamellar graphite with a high proportion of perlite resulting from the primary austenite (fig. 5). The hardness of such a structure is modest (290 Hv).

IV. WELDING WITH FUSIBLE CAST-IRON WIRE-GUIDE

4.1 Problems posed by the use of wire-guides

Wire-guides are simply prepared from two rectangular elements of equal length, whose profiles when placed symmetrically together form a rectangular section having a central channel obtained by machining (fig. 6a).

With a solid wire of mild steel, the section of the wire-guide must satisfy two opposing requirements:

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- it must be as large as possible in order to limit the loss of carbon and silicon from the molten area;
- it must remain sufficiently small to ensure:
  - . suitable electrical insulation,
  - . the introduction of the slag and inoculant alloys,
  - . the evacuation of gaseous reaction products.

Figure 6c which represents the theoretical variation of the carbon concentration as a function of the relative width ($\frac{l}{e}$) of the wire-guide, shows that:

- the carbon saturation index (Sc) is limited to a modest value;
- the increased concentration is not sufficient to justify the use of large section wire-guides.

As far as the grade of the filler wire is concerned, various solutions are open. However, the solid wires used for the welding of non-alloy steels are of an unfortunate composition (low carbon, high manganese and low silicon) as far as the graphitisation of the molten area is concerned.

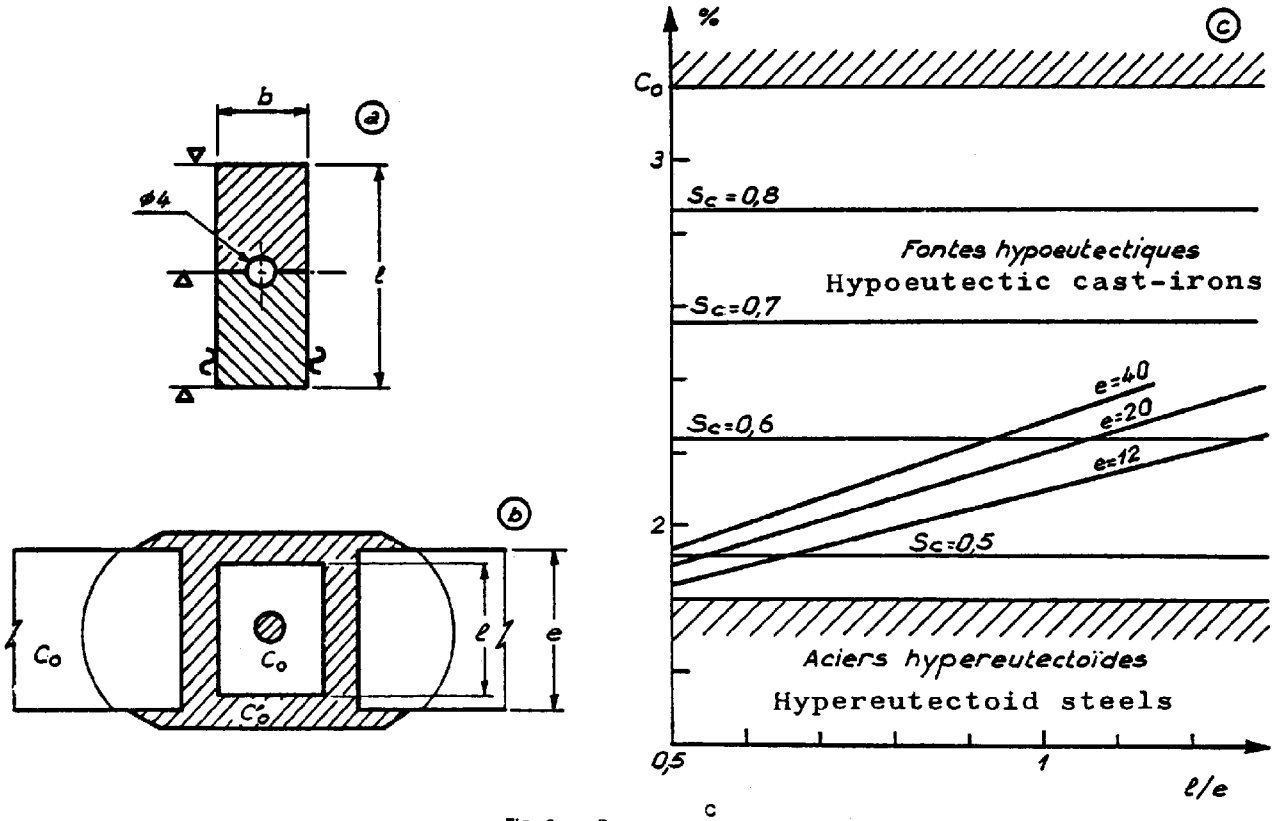


Fig. 6. — $Sc = \frac{C}{4,3 - 0,3\ Si}$

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4.2 Preliminary tests with solid steel wire

Several tests were conducted under different operating conditions, in order to observe the behaviour of the slag with the fusible cast-iron wire-guide technique.

The filler material wire used for this series of tests comprised 1% Mn and 0.20% Si. The support pads were of water-cooled copper.

From the point of view of the soundness of the joints, all the results obtained were high satisfactory. No inclusion was detectable between the molten area and the base metal.

As far as the structure of the molten area was concerned, the 20 mm gauge plates in each case showed a white texture of redhibitory hardness (350 to 480 Hv) principally consisting of perlite and ledeburite (fig. 7).

On the thicker plates of F_T25 cast-iron, the molten area had a mixed structure whose core hardness did not exceed 360 Hv (fig. 8).

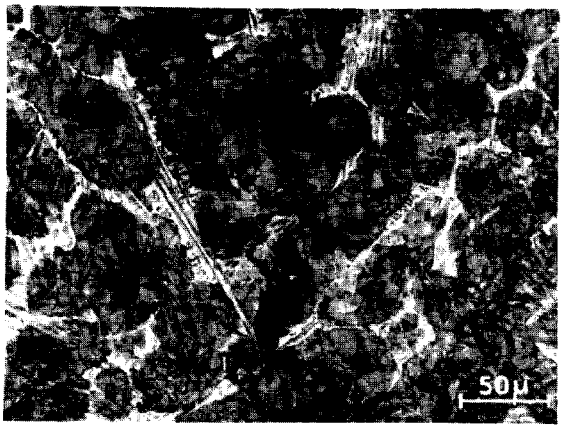


Fig. 7.

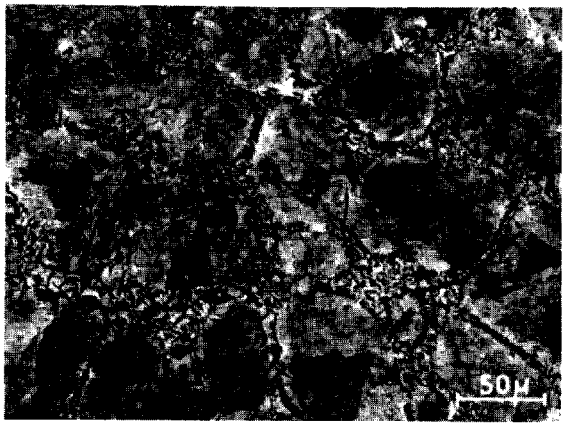


Fig. 8.

4.3 Use of a coated wire depositing a SG cast-iron

The use of a coated wire depositing a spheroidal graphite cast-iron in gas-protected welding seemed appropriate from the standpoint of the graphitisation of the molten area.

Two tests were conducted on these lines, with a coated wire containing a powdery mixture of graphite and of ferro-silicon.

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Each test resulted in failure. The operating conditions become very unstable as soon as the coated wire melts and considerable effervescence is observed in the slag with the practically continuous occurrence of stray arcing.

The molten area then offers the appearance of an irregular ingot separated from the base material by large inclusions of slag.

4.4 Welding tests with graphitising inoculation

The inoculation technique, frequently used in the foundry industry, needed looking into in order to examine its influence on the structure of the molten area.

However, vertical welding under slag does not allow, as far as the inoculation is concerned, the freedom of action available in the foundry; in particular:

- the temperature of the liquid is fixed by that of the slag;
- the inoculant, which has to be introduced as the molten area progresses, is in contact with a reagent slag.

Several inoculant alloys were used during this series of tests (table III). Certain products, which were insufficiently dense or considered too reactive as far as the slag was concerned, were deliberately rejected.

The alloys were first crushed to a suitable granulometry, then gravity-fed during the welding, using a vibrating bowl distributor.

Water-cooled copper support-pads were again used, in order to maintain thermal conditions similar to those observed during the previous tests.

TABLE III

Ref.	Standard composition	Density	Granulometry (mm)
IG1	Fe Si45	4.4	Powder 0.1-0.4
IG2	Fe Si75 Al Ca	2.9	Grains ≤ 1.6
IG3	Fe Si65 Ba Al Ca	3.7	" "
IG4	Fe Si75 Zr Ca Al	3.0	" "
IG5	Fe Si65 Mn Zr Al	3.4	" "
IG6	Fe Si65 Ca30 C	2.5	" "

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The results obtained proved conclusive. The perfection of the graphitisation of the molten area nevertheless seems to be linked to the quantity and to the homogeneity of the inoculant alloy filler material.

At the edges, the forced cooling of the pads is however accompanied by a cementiticborder which cannot be eliminated despite an abundant inoculation of ferro-silicon (fig.9).

The central part of the molten area has a perlite-ferritic structure with fine interdendritic graphite of type D (fig.10).

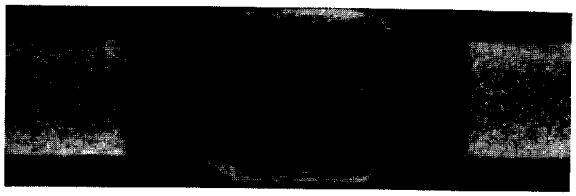


Fig. 9.
F_T25. e = 20 mm. Laitier A. Inoculant IG1, taux : 4,5 %.

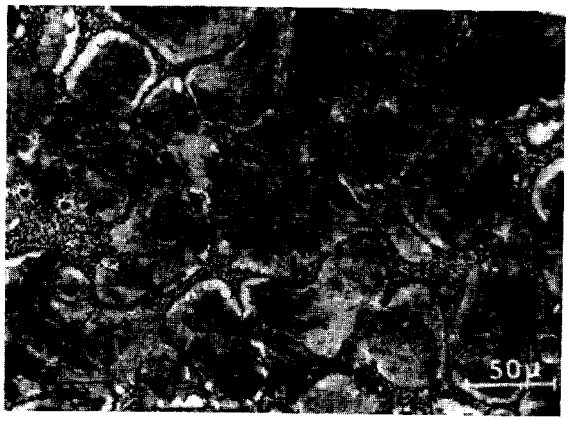


Fig. 10.

Fig. 9
F_T25. g = 20 mm. Slag A.
Inoculant IG1, rate:4.5%

4.5 Use of refractory pads

The use of a refractory material for the support pads is doubly advantageous from the graphitic solidification standpoint:

- decrease in the cooling speed of the molten area, following the increase in welding energy;
- increase in the dilution rate and consequently, in the carbon and silicon content of the molten area.

Several refractory substances were used (table IV). Each of the two pads is placed between two lateral coolers whose purpose is to stabilise the thermal conditions (fig.11).

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TABLE IV

Ref.	Refractory material
R1	Graphited alumina (75%)
R2	Amorphous carbon
R3	Graphite
R4	Mullite ($3\ Al_2O_3,\ 2SiO_2$)
R5	Sintered silica

In general, considerable savings are obtained in the welding energy and the rate of dilution.

During the welding, the behaviour of the refractories differs according to the substance used:

- despite prior stoving because of their porous structure, silica, graphited alumina and mullite release considerable quantities of gas which makes the heat transfer somewhat unstable;
- the amorphous carbon suffers moderate erosion leading to a local carbon enrichment of the molten area, but also a heat cycle sensitivity which causes the pads to crack;
- the graphite offers excellent thermal shock resistance and the erosion of its surface remains negligible if the latter has a machining "polish".

As far as the structure of the molten areas is concerned, the results were fairly disappointing in the main. In particular, the spheriodal graphite cast-iron, even for the thicker plates, showed carburic solidification similar to that observed during the preliminary tests. The lamellar graphite cast-iron, however, offers a mixed structure whose graphitisation appears to be favoured by the use of amorphous carbon pads.

4.6 Welding with refractory pads and inoculation

The simultaneous use of these two techniques enabled us to achieve very satisfactory results.

In addition to gravity inoculation, two other techniques were studied during this series of tests:

- coating of the wire-guide with inoculant alloy granules;

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- feeding of the inoculant powder through the electrode-wire.

Moreover, several operating variables were examined; these involved in particular:

- the use of a small section wire-guide;
- the influence of the nature and granulometry of the inoculant alloy,

Finally, a wire with a lower Mn content (0.4%) was used, with support pads either of graphite or of amorphous carbon.

From the point of view of the inoculation technique, it appears that:

- coating of the wire-guide offers a suitable homogeneity and rate of inoculation;
- the feeding of the powder through the wires does not allow a sufficient rate of inoculation to be obtained;
- gravity inoculation remains somewhat imprecise and requires prior calibration of the feed device.

From the standpoint of the graphitisation efficiency, the nature and granulometry of the inoculant alloy cause no major difference to appear.

A large amount of inoculation (3%) moreover ensures sufficient graphitisation of a little diluted molten area, obtained using a small section wire-guide.

By contrast, even massive inoculation does not allow the local carburic hardening to be eliminated, which appears on thin plates (12 and 20 mm) in the areas beneath the lateral coolers.

In general, graphitisation is favoured by increasing the thickness of the plates. Figure 12, for a 40 mm thick F_T25 plate, shows abundant lamellar graphite and a high proportion of ferrite. The hardness of such a structure is modest.

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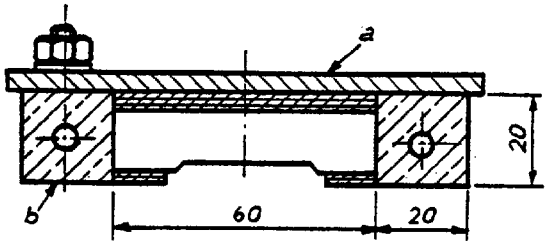


Fig. 11.
a : amiante ; b : refroidisseur cuivre.

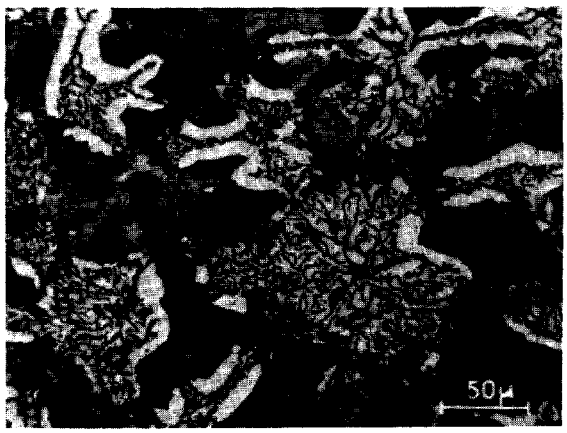


Fig. 12.

a: asbestos; b: copper cooler.

Figures 13a and 13b are a schematic representation of the structures of the molten areas observed for each of the grades and thicknesses studied.

4.7 Attempts to nodularise the graphite of the molten area

The positive results obtained at the outcome of the preceding tests naturally led us to study the possibility of nodularising the graphite of the molten area of the plates of F GS 50 cast-iron.

Two channels were explored:
- use of a coated wire comprising a ferro-silico-mischmetal powder;
- gravity inoculation with magnesium alloys.

Despite stable operating conditions, each of the tests performed produced more or less abundant inclusions of slag in the molten area.

From the metallography standpoint, the object appears partially achieved by the coated wire technique; the central part of the molten area in fact shows a nodular graphite of small dimensions, with a halo of ferrite and finely divided in a perlitic structure (fig. 14).

By contrast, and despite the massive introduction of magnesium alloys, the nodularising inoculation proves totally inaffective. There are grounds for thinking that the magnesium introduced reacts with certain oxides of the slag.

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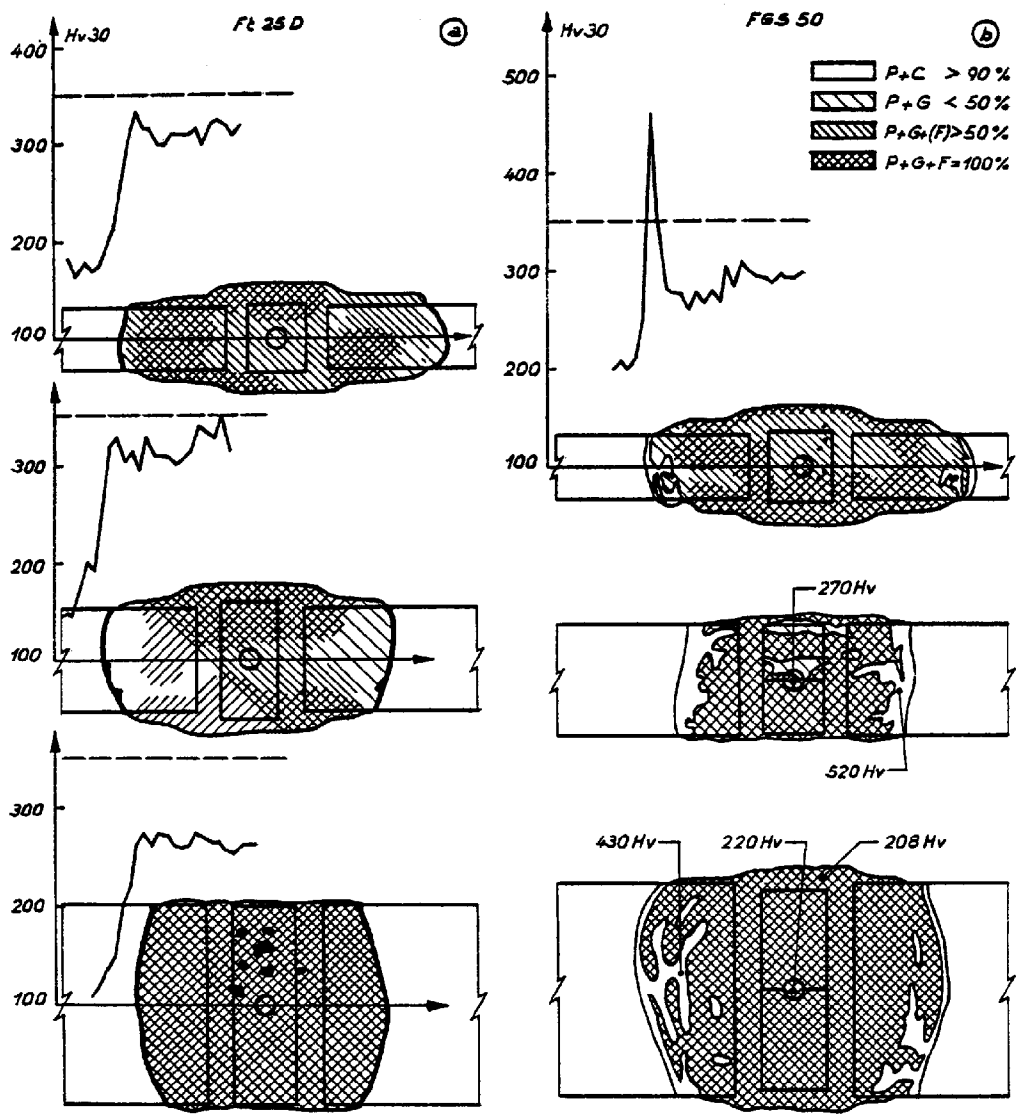


Fig. 13.

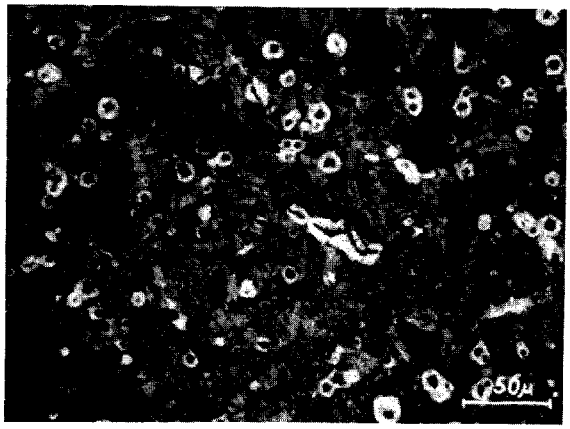


Fig.14
FGS50. g = 20mm. Coated wire
Fe-Si-Ce

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V. STUDY OF THE BEHAVIOUR OF THE SLAG

The tests described above show profound differences in the behaviour of the slag. Various examinations were conducted on the residual slags and on the inclusions observed during certain tests.

5.1 Welding with cast-iron rectangular rod

One of the inclusions observed was subjected to a metallographic examination using a longitudinal section taken from the initiation area. The welding sample used for this examination comprised an F_T25 cast-iron plate and an E24-1 mild steel plate (fig. 15A).

The micrograph in fig. 15B shows local carburization of the edge of the steel plate and the presence of an intermetallic compound (Fe-Si-Mn-C) between the slag and the previously carburized area.

These observations are the result of the presence in the slag of carburized compounds (CO, CSi, C_2Ca, Fe_3C) caused by the oxido-reduction reactions at the rod-slag interface between the graphite and the various oxides (fig. 16). The effervescence resulting from these reactions causes fluctuation of the electrical contact with a corresponding drop in the rms current and in the temperature of the layer of slag. The electrical oscillograms in figures 17 illustrate the effervescence phenomena observed; the marked instability visible in figure 17b reveals an intense formation of CO as a result of the low dissolution speed of the lamellar graphite. As the viscosity and surface tension are considerably influenced by the temperature, one may imagine the formation of a viscous layer which works against the bonding of the deposited metal.

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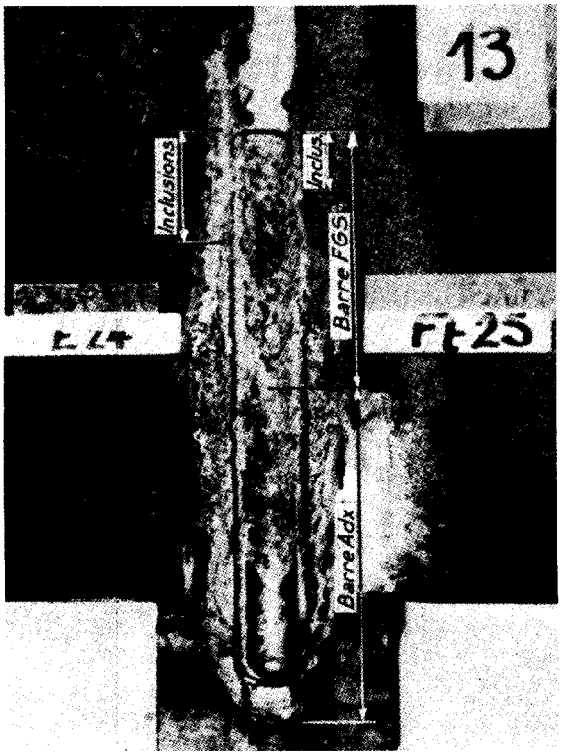




Fig. 15.
15 A (à gauche).
15 B (ci-dessus).

Fig. 15
15 A (left)
15 B (above)

a: carburized area
b: intermetallic compound
c: slag

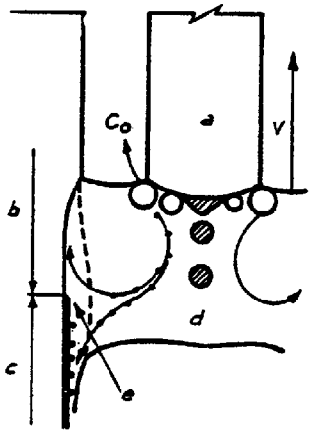


Fig. 16.
a : barre fonte.
b : cémentation.
c : composé intermétallique.
d : laitier.
e : couche visqueuse.

$$FeO + C \rightleftharpoons Fe + CO \nearrow$$
$$MnO + C \rightleftharpoons Mn + CO \nearrow$$
$$SiO_2 + 2C \rightleftharpoons Si + 2CO \nearrow$$
$$SiO_2 + 3C \rightleftharpoons CSi + 2CO \nearrow$$
$$CaO + 3C \rightleftharpoons C_2Ca + CO \nearrow$$

a: cast-iron rod
b: case-hardening
c: intermetallic compound
d: slag
e: viscous layer

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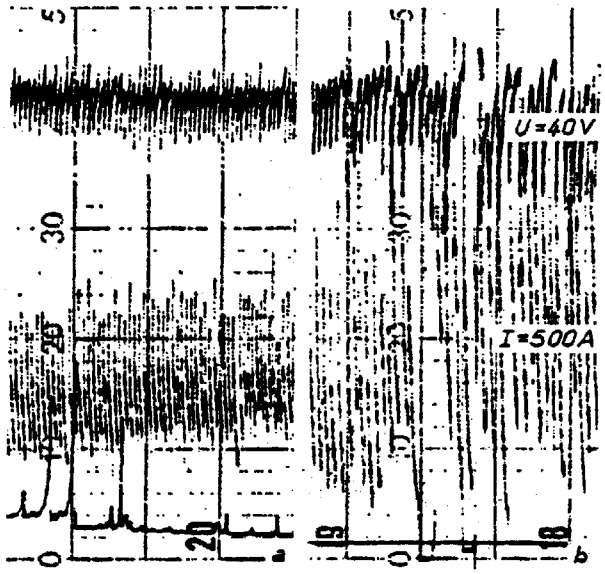


Fig. 17.
17a : barre FGS. Laitier A : 17b : barre F_T25. Laitier A.

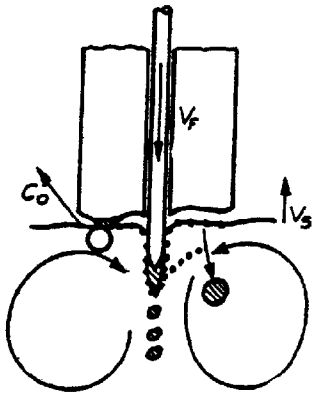


Fig. 18.

Fig. 17
17a : FGS rod. Slag A;
17b: F_T25 rod. Slag A.

5.2 Welding with cast-iron wire guide and mild steel wire

The penetration of the wire into the slag ensures a regular electrical contact free of intermittent effervescence reactions (fig. 18). Moreover, the carburized particles are rapidly brought to the surface of the wire by the convection movements and case-harden the latter.

5.3 Welding with coated wire depositing a cast-iron

Because of the high speed of the wire, a considerable part of the graphite is in contact with the slag and this results in abundant effervescence which acts as an electrical screen.

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VI STUDY OF STRUCTURES

6.1 Structure of the heat affected zone and of the bonding zone (figure 19)

Below solidification temperature (approx. 1180°C) F_T25 lamellar graphite and spheriodal graphite cast-irons behave in an identical manner. Transformation occurs at around 740°C, and fragmentation of the perlite is observed following the local austenitisation.

With the rise in temperature, the carbon concentration of the austenitic matrix increases under the effect of the diffusion of the secondary graphite. Because of the low cooling speed, the perlitic transformation is assured, and the local diffusion of the carbon results in the formation of rings of ferrite around the initial graphite lamella.

Within the fusion interval (1180-1230°C), the two grades behave differently. While the F_T25 cast-iron causes a perlite-ferritic structure to appear with fine graphite lamella, the residual magnesium of the F GS 50 cast-iron causes a carburic solidification accompanied by further nodularisation.

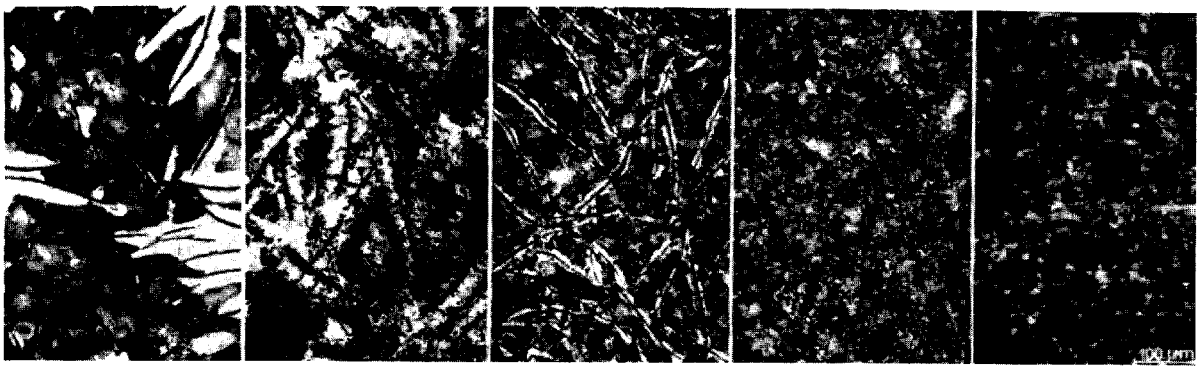


Fig. 19 a.

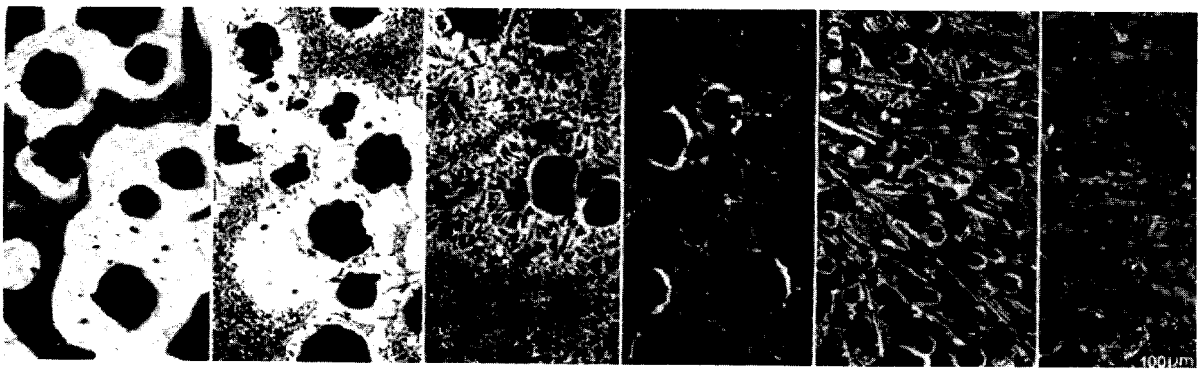


Fig. 19 b.

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6.2 Structures of the molten zones

Certain parameters on which the structure of an as-solidified cast-iron depends are determined by the welding operation. Thus:

- the chemical composition of the molten "bath" depends in the first place upon that of the basic cast-iron because of the dilution, and in the second place upon the nature of the slag;
- the cooling speed depends upon the welding energy used, which in turn depends upon the thickness of the parts;
- the temperature of the molten "bath" is determined by that of the layer of slag.

6.2.1 Graphitisation - Graphitisation potential

At the eutectic temperature, the graphitisation reaction: $Fe_3C + FeSi \rightarrow CSi + 4\ Fe \rightarrow 3\ Fe + FeSi + C$ (graphite), is governed by three basic factors:

- cooling speed of the liquid;
- temperature difference of the eutectic levels of the Fe-graphite and Fe-Fe 3C systems;
- presence of crystallisation grains in the liquid.

The latter two factors can each be expressed in the form of a function of the carbon and silicon concentration:

$$\sigma_1 = f\ (C,\ Si)$$
$$\sigma_2 = g\ (C, Si)$$

The graphitisation potential can thus be defined by $P_G = \sigma_1 . \sigma_2$, which gives us an equation of the type:

$$P_G = \frac{Si(C - 2.03 + 0.15\ Si)}{2.27 - 0.15\ Si}$$

6.2.2. Influence of inoculation

Chemical analysis of the molten zones allows us to show in a diagram (C, Si) the dots representing the structures observed during the various tests (fig. 20).

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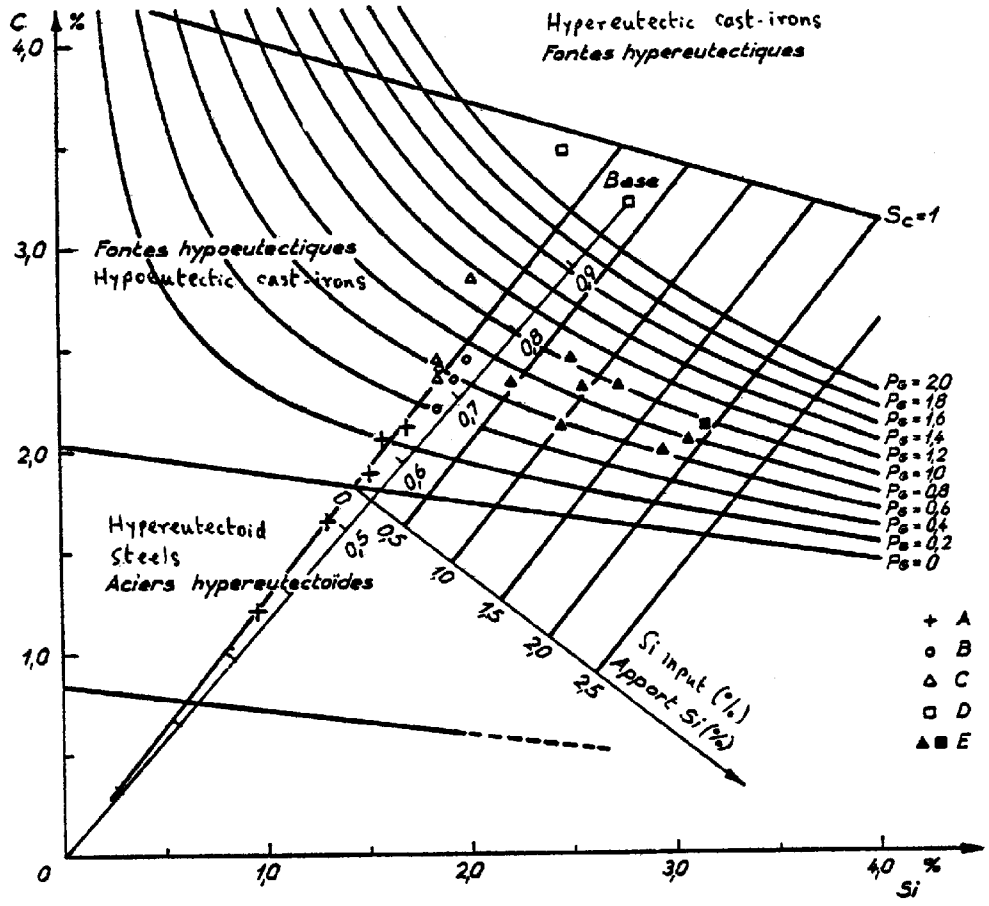


Fig. 20. — Fonte F_T25. Epaisseur 20 mm.

Fig. 20 - F_T25 Cast-iron. Thickness 20 mm

Ref.	Structure	Components	Approximate concentration
A	White	Perlite + cementite	P + C > 90%
B	Mixed	Perlite + cementite + graphite (spots)	P + G < 50%
C		Perlite + graphite (D or E) + cementite (spots)	P + G > 50%
D	Black or grey	Perlite + graphite (D or E) + (ferrite)	P + G + (F) = 100%

E : inoculated tests.

In the absence of inoculation, the dots are roughly aligned in a straight line whose slight shift compared with that joining the

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origin to the base composition, reveals a loss of Si. In the case of a 20 mm thick F_T25 cast-iron, suitable graphitisation would require a P_G of at least 0.8, which in practice rules out the fusible wire-guide and mild steel wire technique.

On the other hand, inoculation allows the use of this technique, by lowering the graphitisation threshold.

6.2.3. Influence of energy and of the grade

Figures 21a and 21b represent, as a function of the welding energy and of the graphitisation potential, the structure and hardness observed in the centre of the molten zone for each of the two grades studied with the fusible wire-guide technique.

By contrast with grade F_T25, for which the component elements and hardness develop progressively in the non-inoculation and inoculation areas, the F GS 50 cast-iron causes two distinct structures to appear separated by an insufficient inoculation area where carburic and graphitic solidifications co-exist.

The particular sensitivity of GS cast-iron to metastable hardening must be attributed to the residual Mg whose "whitening" power is well known.

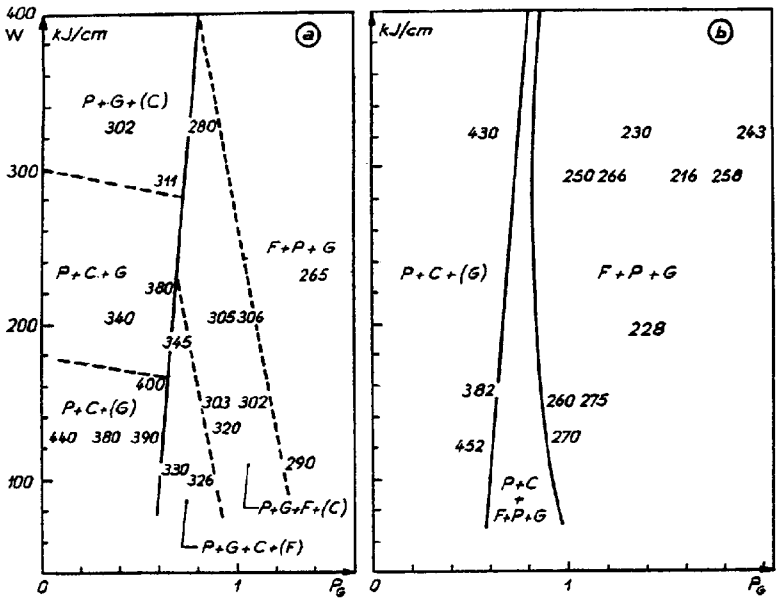
6.2.4. Heterogeneity of the molten zones

An example of prismatic rod welding is particularly well represented by the macrograph in figure 2b, where carburic areas lie alongside highly ferritic graphitic areas, despite the absence of specific inoculation.

There is reason to believe that the molten cast-iron receives a local contribution of carburised particles deposited in clusters as a result of the convection movements, thus ensuring an involuntary heterogeneous inoculation.

With the exception of the local hardening caused by the unfortunately necessary presence of the lateral coolers, wire-guide welding is generally characterised by a sufficient homogeneity of the molten zone. However, despite massive inoculation, F GS 50 cast-iron causes the appearance of spots whose structure is identical to that observed in the bonding zones (fig. 22a). The longitudinal fractography in figure 22b shows that these are in fact "scraps" of

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low concentration. Fig. 21.
[ ] : faible concentration. 21a : F_n25 ; 21b : FGS50.
P : perlite ; G : graphite ; C : cémentite ; F : ferrite.

non-miscible base metal formed by the mechanism illustrated in figure 22c. The lack of miscibility observed is the result of the alteration caused by the MgO to the surface tension of the liquid GS cast-iron.



Fig. 22 a.

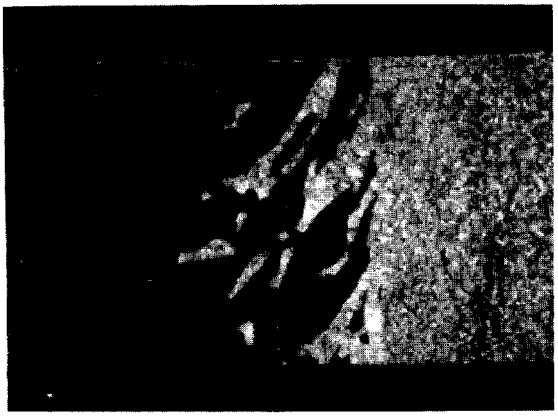


Fig. 22 b.

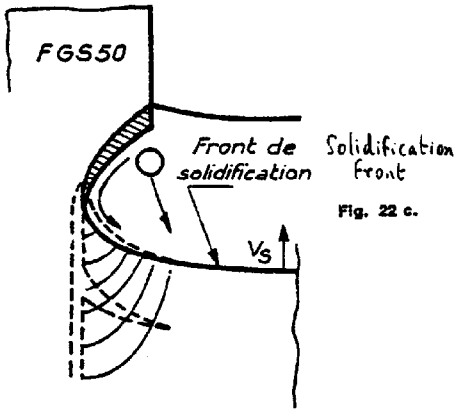


Fig. 22 c.

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VII CONCLUSIONS

The main objective of the study has been achieved on general lines; thus, after steel fabrication, followed by light alloy fabrication, vertical welding techniques under slag can now be extended to cast-iron fabrication.

From the operating method standpoint, it seems useful to insist upon the simplicity of the fusible wire-guide technique. Moreover, the extension of the welding heat cycle allows the user to dispense with the delicate and troublesome pre-heating operation, which is essential with the traditional so-called "hot thermal condition" processes.

From the metallurgical angle the results obtained differ considerably according to the grade of cast-iron.

- Lamellar graphite cast-iron in the end offers excellent weldability; the mechanical strength of the joints formed is greater than that of the base metal, while remaining perfectly machinable.

- For its part, spheroidal graphite cast-iron is characterised by carburic hardening of the bonding area, which may present problems in the rough-welded state, as far as further machining is concerned, together with risks of transverse cracking. Moreover, the lamellar graphitisation of the molten zone does not give the joint the good mechanical characteristics of the base material and the spheroidisation of the graphite in the molten zone remains a problem.

SUMMARY OF THE DISCUSSION FOLLOWING M. MICHEL'S LECTURE

M. Granjon: Before putting a question to M. Michel, I would like to point out, because I feel myself somewhat responsible, that M. Michel's work involved an exceptional number of failures before satisfactory results were finally obtained, and that he overcame a series of obstacles rarely encountered in research work.

Having said this, I shall now put two questions to M. Michel:

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I note that the problem of welding lamellar graphite cast-iron has practically been solved, which is not the case with spheroidal graphite cast-irons. However, you have only studied rough welds; would it not be interesting to carry out referritisation treatments upon the spheriodal graphite cast-irons and even upon the lamellar graphite cast-irons?

M. Michel: Yes, it would be interesting to study this parameter.

M. Granjon: In your conclusion, you said that your process would make it possible to weld iron castings and moreover that it was possible to carry out heterogenous welds such as steel to cast-iron; this latter point seems to me to be an interesting development. What do you think?

M. Michel: The heterogeneous welding of steel to cast-iron is indeed possible and we then find ourselves faced with a choice of structure of the molten zone: either mild steel or cast-iron. I think that the latter would be preferable.

M. Margerie: I think that indeed for spheriodal graphite cast-iron, particularly where it has remained carburic, annealing should make it possible to quickly and easily return to a compact graphite structure; if lamellar graphite has already been precipitated, it will remain and it will not be possible to have very high ductility characteristics; as far as hypereutectoid steel is concerned, this is not my speciality and I believe that it is fairly difficult to graphitise by heat treatment.

M. Lieger: Are the mechanical and structural characteristics of the cast-iron better in the case of welding of thicker parts, 100 mm for example?

M. Michel: As you will have seen, the thicker the cast-iron, the better the results, because the cooling speed is lower and graphitisation is that much easier. We limited ourselves to thickness of 12 to 40 mm, because our test equipment did not allow us to study greater tnicknesses.

M. Lieger: We the users and workshop people are not as optimistic as you; some of the applications seem relatively tricky.

M. Michel: Clearly vertical welding under slag only applies to rectilinear welds and to gradual thickness variations.

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M. Decrop: In view of the importance of the inoculation treatment, according to the tests which you have done, on the very marked structural changes obtained, one may ask if in the case of spheriodal graphite cast-irons, which appeared the most difficult, it would not be best to improve the effectiveness of the inoculation treatment by altering the quantity or quality; you always used 75% ferro-silicon, but additives can be added to this product such as aluminium, calcium or barium, for example.

M. Michel: I do not believe that studying other inoculant products would enable us to eliminate the carburic solidification which occurs in the bonding zone; this is a zone which is unaffected by the inoculation, and in any case we would be left with local carburic hardening for which nothing apart from further heat treatment could be done.

M. Decrop: Since this zone assumes a carburic structure, this means that it has passed through the liquid state, and therefore could be sujected to inoculation.

M. Michel: No - according to the heterogenity study, there would be a miscibility gap.

M. Decrop: The temperature would be too low for the dilution of the inoculant to occur.

M. Dejieux: You have often said that when welding with prismatic rods the graphite is a problem; did you ever try using white iron rods?

M. Michel: Yes - we thought about this solution. It is true that in this way the effervescence reactions would not have occurred at the rod/slag interface, but unfortunately we found that it would be very difficult technically to produce white iron rods which were sufficiently accurate in size and straight; however I am convinced that this solution would have given good results.

M. Margerie: Coming back to M. Decrop's idea of increasing the effectiveness of the inoculation, we might think about graphite inoculants; while it is clear from what you have said that this type of inoculant is ruled out because of the reactions which may take place with the slag, for the other special inoculants with barium, strontium and zirconium on which I believe you have carried out a number of tests, are the differences great?

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M. Michel: Calcium and graphite-based inoculants were automatically ruled out, and we had problems with a calcium ferrosilicon; the presence of calcium altered certain physical properties of the slag and caused inclusions. The best results were obtained with ferro-silicon with a high silicon content and possibly barium and aluminium additives.

M. Decrop: You briefly alluded to the mechanical characteristics of the welded samples; can you give us a few indications?

M. Michel: We carried out transverse rectangular tensile tests on joints 40 mm thick, on the one hand in the case of a lamellar cast-iron (Ft 25) and on one the other hand in the case of spheriodal graphite cast-iron (FGS 50); in the first case, rupture occurred outside the weld for a breaking load corresponding to that of the base metal, and in the case of the spheriodal graphite-cast-iron the rupture occurred in the weld, the strength of the molten zone being lower than that of the base metal.

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Flash Butt Welding of Cast Irons to Other Metals

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This paper was originally presented at the 44th International Foundry Congress in Florence, Italy in September, 1977 and is translated and reprinted here in full.

ABSTRACT

This paper describes the flash butt welding process as it is used to join cast irons to other ferrous and nonferrous metals, including steels, copper, brass and aluminum. Post-welding heat treatments and the metallurgical characteristics of the welds are discussed. Particular applications of this welding technique to industrial production are considered.

Introduction

Electric resistance welding is a method of autogenous welding by pressure and without adding metal, where the heat necessary to bring about local preheating of the surfaces to be welded is supplied by electrical resistance by passing a current across the surfaces which are in contact at the weld zone. This procedure requires an electric current at very low voltage, on the order of 4 to 8 volts, for most applications, or from 1 to 30 volts when it is desired to cover the entire range of possibilities, but at an extremely high current intensity, usually on the order of several hundred amperes, but which can even reach more than 100,000 A when dealing with very heavy welds. The value of the current intensity varies according to the type of metal to be welded, the welding time and the pressure used. All these parameters are controlled by means of special electronic controls on the welding machines.

Electrical current of the correct voltage and amperage is supplied by transformers built into the welding machine. Current from these transformers goes to the contact electrodes to the surfaces to be welded, which also form part of the electric circuit. The energy finds the greatest resistance to its flow at the contact point of the two metallic surfaces, which are either face-to-face or leaning (abutting) one against the other, and heats them by ohmic resistance in such a manner that they are quickly brought to the preheated (sub-fusion) state. The heat given off in the welding zone or through the electrodes is expressed in kilocalories by the formula:

$$Q = \frac{I}{J}(I^2RT)$$

where I is the current intensity in A, R is the sum of electrical contact resistances between the surfaces to be welded (measured between the electrodes), T is the time of passage of the current, expressed in sec, and J is the mechanical equivalent of the calorie, 1 cal = 4.18×10^3 J. The automatic application of sufficient pressure to the electrodes during or after the time of current flow insures a good weld junction.

Butt welding takes place on the entire frontal surface of the parts to be welded, which are placed face-to-face and squeezed at high pressure between the current-carrying jaws of the welding machine (Fig. 1). These jaws are of a metal which is a good electrical conductor and the surfaces of the parts in contact with the jaws should be clean and bright, not oxidized, to insure the best possible conductive path for the current. The welding cycle is shown schematically in Fig. 2. In the case of flash butt welding, one of the two jaws is fixed, the other movable; this latter advances toward the former in such a way that the two faces to be welded approach each other, touching at the tips of their rough surfaces, making it unnecessary for these surfaces to be ground down or parallel.

Upon applying the current, the small points in light contact with each other melt, and the droplets of metal explode and are projected outward. Thus, between the contact surfaces, small electric arcs are formed which propagate over the entire width of the surfaces and cover the surfaces themselves with a film of liquid metal. When a sufficient temperature is attained, a strong sudden pressure is automatically applied to the parts and, at the same time, the current is cut off. In this way, the liquid film, with all the impurities it contains, is squeezed out and the weld comes about by recrystallization.

Source: AFS International Cast Metals Journal, June 1979 (American Foundrymen's Society)

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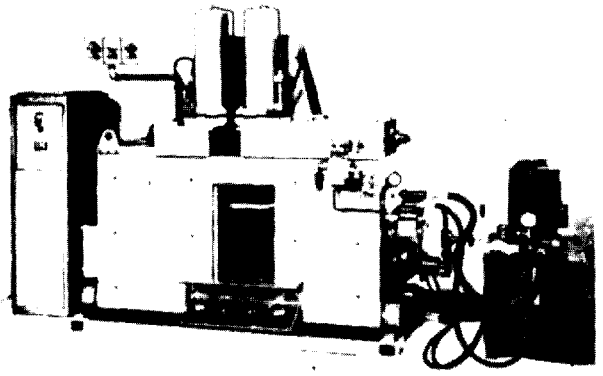


Fig. 1. Test apparatus.

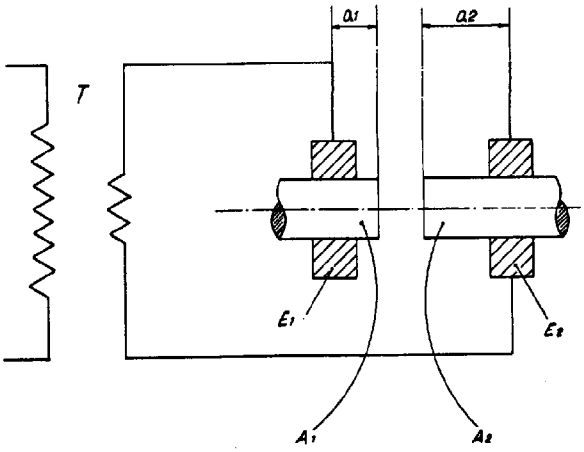


Fig. 2. The two parts to be welded, A1 and A2, are held by two electrodes cooled by water, E1 and E2. These stick out from the part a certain predetermined distance, a1 and a2. One of the electrodes is movable.

This welding procedure is widely used for joining steel and nonferrous metals and their alloys. Automatic machines are available which, in spite of their high purchase price, use relatively little energy and are well suited to mass production in such industries as automotive and aircraft manufacture, transport vehicles, electrical kitchen equipment, hardware, cooking utensils, metal furniture, and in the joining of rails and water and gas mains.

Flash Welding of Cast Iron

Cast iron, especially gray cast iron, is considered a difficult material to weld in the strict sense which is given to this term when comapred with soft steels. Actually, all types of cast iron constitute a category of metals which is not easy to weld because of their high carbon content, whether free or combined, in the form of graphite having different morphologies, the presence of which assures that the metal has a very low capacity for supporting thermal shocks and is easy to break. The capacity for enduring the thermal shocks which take place during the welding cycle is considerably affected by the morphological structure of segregated graphite and, in general, nodular graphite irons and floccular graphite irons (ductile and malleable irons) withstand these shocks better than flake graphite irons (gray irons in general).

The first tests for flash welding of cast iron were carried out in Russia in 1936 and they were extended in the following years, especially at the Paton Institute of Kiev which, in 1958, announced the good results which could be achieved with this system of welding and the considerable advantages it presented with respect to other welding systems.

These advantages are:

1) Heating is uniform and simultaneous throughout all sections to be welded.
2) Welding is carried out very rapidly, so that the influence of temperature on the base metal is very limited and the risk of residual stresses is reduced.
3) Because of the presence of free graphite segregation, the cast irons should be classed among metals characterized by poor miscibility in the solid state. The procedure using flash welding is, in that case, the most suitable, since it is a method of welding without adding feed metal. The welding is performed by forced interpenetration of metal atoms from the surface to be welded brought to the plastic state by preheating, while maintaining absolute continuity of the metal.
4) The extremely numerous electric arcs which form during the flash welding phase, by carrying a certain layer of material from the surface, accomplish perfect cleaning of the latter of all traces of oxides, inclusions and graphite segregation.
5) The cast metal, rich in heterogeneous nonmetallic materials, is entirely squeezed out of the joint during the compression phase and it solidifies on the outside in the form of a bead. As a consequence, the welded metal is metal which has not been superheated above the melting temperature, thereby reducing the possibility of forming hard, fragile carbide structures in the weld zone.
6) During welding, and precisely during the flashing phase, the stratum of liquid metal at the surface is highly superheated, with the consequent combustion of carbon and decarburization of the surfaces which are face-to-face for welding, as well as the contiguous zones, therefore favoring carbon diffusion and partial diffusion of the segregated graphite..

On the other hand, the difficulty encountered in flash welding of cast iron is caused by the low ductility of the metal. Cast iron has, actually, a very low capacity for hot deformation. This is why the squeezing phase should be carried out with the greatest care, in such a way that all the molten materials are squeezed out of the weld zone, but in such a way that the base metal situated behind the joint zone is not deformed, because this can cause cracks.

Metallurgy of the Welded Joint

In the course of the welding cycle, as it has been briefly described above, the parts to be joined are heated to various temperatures which gradually decrease, starting from the zone of the joint and the area of surface contact, where the melting temperature of the metal is reached, down to temperatures slightly higher than ambient at the back of the jaw clamps.

However, a certain increase in temperature is observed throughout the entire welded part, but only as a consequence of the thermal conductivity of the metal. Consequently, even if the different zones of the welded part are submitted to these thermal influences for extremely brief times, structural transformations take place within the metal. The transformations are linked to carbon diffusion phenomena and these depend on the level of temperature reached in the different zones, on the time these temperatures are maintained, and on the speed of cooling which follows.

The diagram of Fig. 3 is a schematic representation of the possible structural variations noted over the length of cast iron

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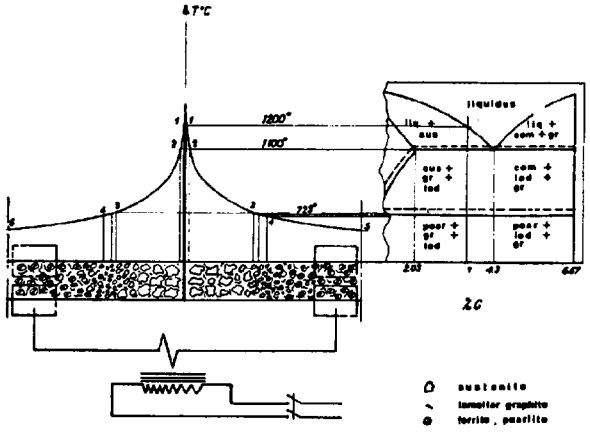


Fig. 3. Structural and Crystallographic Modifications.

Part	Zone	Modification
C-1	Superheat zone	Total or partial solution of graphite in austenite.
1-2	Melt zone	Austenite and graphite not dissolved. The dimension of the austenitic grain increases as a function of increasing temperature and the dimensions of the undissolved graphite particles decrease as a function of increasing temperature. The dimensions of the graphite particles precipitated from the austenite during cooling are more significant than those of the original graphite particles.
3-4	Transition zone	Austenite, ferrite and graphite. A part of the matrix is transformed into austenite, another part remains untransformed. The original pearlite is austenitized and the austenite envelops the graphite particles at the same time that these graphite particles are combining with the iron to give rise to austenite.
4-5	Zone of base metal thermally unaffected	Original ferrite, graphite and pearlite. Spheroidization is possible if the pearlite is artificially aged.

parts during the welding cycle, in which the cooling curve for a cast iron of hypoeutectic composition is compared with the curve for the temperatures developed over the length of the part. On the x-axis, corresponding to the longitudinal section of the part, can be seen the progression of structures which may be formed during flash welding under routine conditions, starting from the joint zone down to the base metal.

The diagram is completed by a brief commentary on the above-mentioned structures relative to a flake graphite iron.

As for the final structure, one must always take into account the cooling rate after welding, which has a considerable influence on its formation.

Observation of the diagram shows that during the welding cycle the matrix in the thermally-influenced zone consists of pearlite and a certain amount of ferrite, besides graphite, and eventually steadite according to the type of cast iron, which is more or less entirely transformed into austenite. When followed by an especially rapid cooling, this gives rise to a pearlitic structure more or less fine, or even to a sorbitic or troostitic structure which is then harder and stronger than that of the original base metal.

At the same time, the amount of free ferrite is lowered and, when the cooling is particularly rapid, small quantities of ledeburite appear, even at the furthermost layers of the parts. At a short distance from the welded surface, there are even martensite formations.

To form gray iron with graphite requires an elevated temperature and a certain period of time. This can be accomplished in a limited measure only during the welding cycle. However, when the silicon (Si) content of the cast iron is especially high, this facilitates, in the hot phase, the solution of carbon in the austenite and, in the following cold phase, it allows separation of graphite surrounded by ferrite formations. On the contrary, in the presence of a low Si content, one may always observe the solution of graphite during heating of the parts. During cooling though, there is no observation of strong graphite segregation and the matrix will be constituted of fine pearlite flakes (sorbite-troostite) in the presence of martensite formations, the formation of these being a function of the cooling rate. In conclusion, the final structures are a function of the chemical composition of the metal and the thermal cycle to which the base metal has been subjected.

It is possible to avoid the formation of hard structures, whether of carbides or quenched structures, by artificially reducing the cooling rate of the joint after carrying out the weld, for example, by means of a post-treatment which can be performed on the welding machine itself. This increases the working time and consequently reduces the production capacity.

The presence of steadite (iron phosphide eutectic) in the base iron does not involve any inconvenience, since it has a melting temperature about 200° C (360° F) lower than that of the metal, and is thus squeezed out of the joint before solidification, so that the presence of steadite in the joint is less important than the composition of the base metal.

Test Welds Between Ferrous Base Metals

Welds between homogeneous cast irons

1) Flake graphite gray iron type G 20 - UNI 15007/69
2) Malleable iron type GMN 35 UNI 3779/69
3) Malleable iron type GMN 45 UNI 3779/69
4) Malleable iron type GMN 65 UNI 3779/69
5) Ductile iron type GS 55/5 UNI 4554
6) Ductile alloy iron resistant to corrosion

Welds between heterogeneous ferrous metals

1) Flake graphite gray iron type G 20 with:
   a) Ductile iron type GS 55
   b) Malleable iron type GMN 35
   c) Steel type Fe45 UNI 3596
2) Ductile iron type GS 55 with:
   a) Malleable iron type GMN 35
   b) Steel type Fe45
3) Malleable iron type GMN 35 with:
   a) Steel type Fe45

The specimens submitted for weld testing were obtained in the following manner:

a) For the gray iron with flake graphite, cylindrical test bars of 30 ± 2 mm diameter were cast in a bonded sand mold with dimensions as shown in Fig. 4.

b) For ductile iron and malleable iron, the test bars were obtained from specimens cast having the shape shown in Fig. 5.

c) For steel the test bars were drawn from rolled bar stock with an as-rolled diameter of 30 mm.

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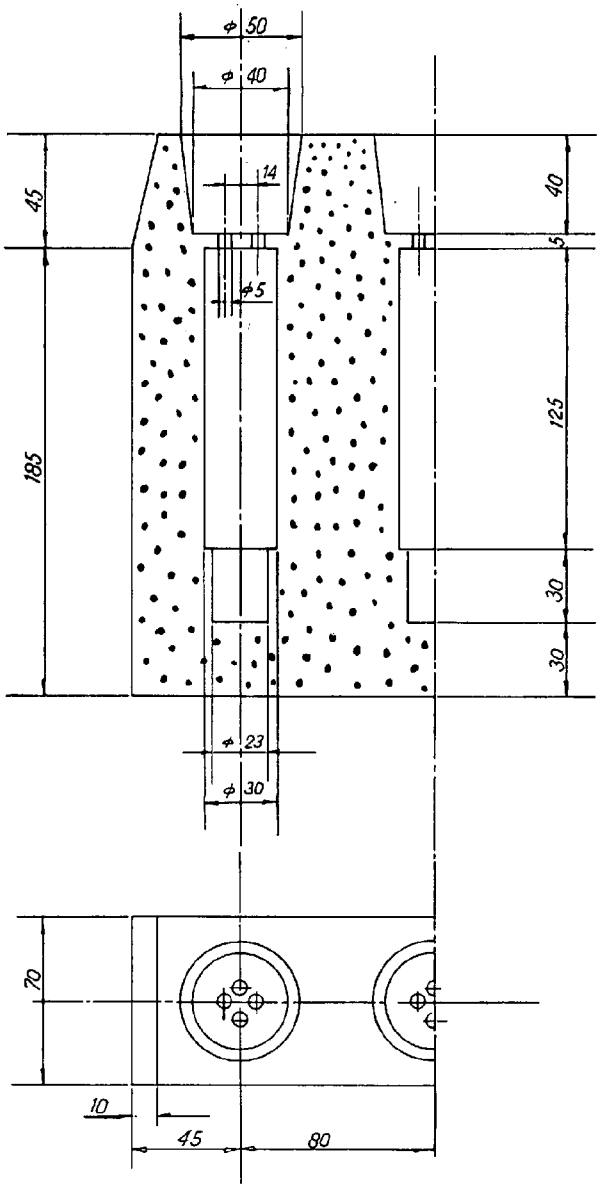


Fig. 4. Dimensions of mold for casting test bars.

The cast specimens shown in Fig. 4 and 5 were retouched on the lathe to obtain cylindrical test bars 25 mm in diameter which were butt welded to metals of the same type and of different types by the flash welding procedure, on the automatic welding machine shown in Fig. 1, which has the following characteristics:

- maximum welding capacity of 600 mm^2
- welding time: 2 to 3 sec
- nominal power at 50%: 140 kVA
- maximum distance between jaws before welding: 40 mm.

This is a production-type machine actually designed for another kind of work: welding of aluminum parts on radiators for thermosiphons.

The welding parameters were adapted, each in its turn, to the type of metal welded, depending on whether this involved welding between two ferrous metals alone, or between non-ferrous metals.

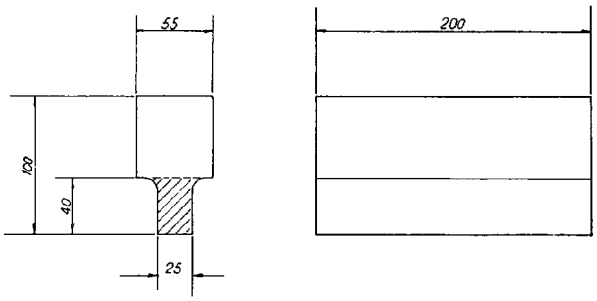


Fig. 5. Test bar dimensions.

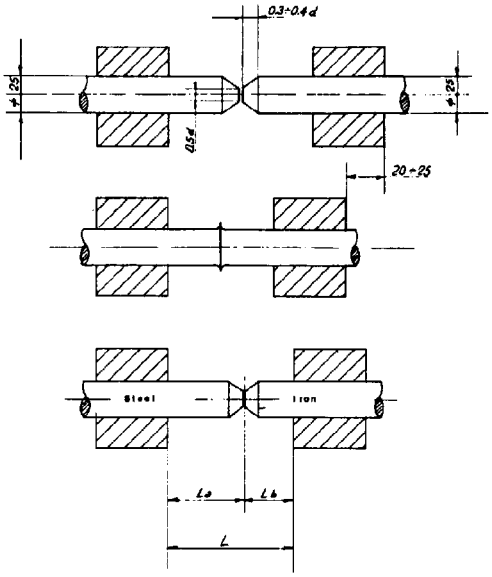


Fig. 6. Positioning and preparation of parts for welding.

The parameters which were controlled in the different welds are the following: squeeze pressure, weld time, feed stress, that is, secondary stress, and preparation of test bars.

Figure 6 shows the positioning and preparation of the test bars, which were the same for all ferrous metals.

The difference between the electrical parameters for the different types of cast iron is not important. A slight difference was noted during welding, especially during flashing, which increased for cast iron with flake graphite a little more than for the other types; on the other hand, the progressive rate, especially toward the end of flashing, should be slightly higher than the parameters usually adopted for welding steel. In the case of welds between different types of cast iron with steel, the extension of the steel test bar with respect to the jaw (Fig. 6) has been modified so as to even up the temperatures of the two parts of the different materials to be joined.

Figure 7 shows the welding cycle and lists the parameters which comprise it.

Post-Treatment of Welded Test Bars, Ferrous Base Metals

In order to verify the structural transformations and the consequent modification of mechanical properties related to the thermal cycle to which the test bars are submitted during and

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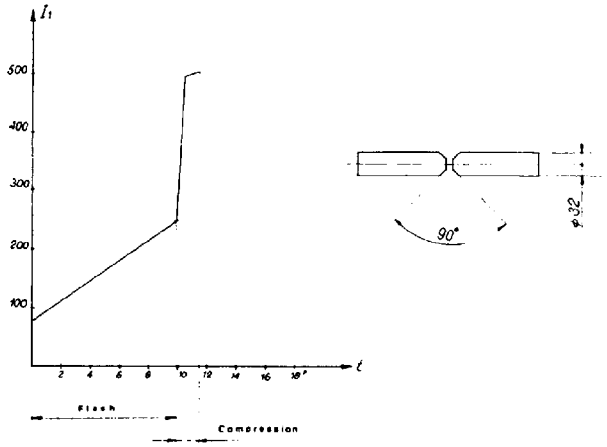


Fig. 7. Welding cycle. t = welding time in sec. I_1 = primary current in amp. Gray iron parts. Squeezing force of 7 kg/cm².

after welding, the cast iron test bars themselves were subjected to different cooling cycles after welding. Some test bars were allowed to cool in air on the machine; others were removed hot from the machine to allow them to cool more slowly on a bed of ashes; still others, after cooling in air, were reheated by electrical resistance on the machine itself, by means of current impulses, up to a temperature of 600 to 650C (1112 to 1202F) for a duration of 50 to 60 sec and then they were allowed to cool.

Tests on Welded Bars of Ferrous Base Metals

Four types of tests were made on the welded test bars of ferrous base metals to determine the following:

- tensile strength
- impact strength
- hardness
- metallographic characteristics.

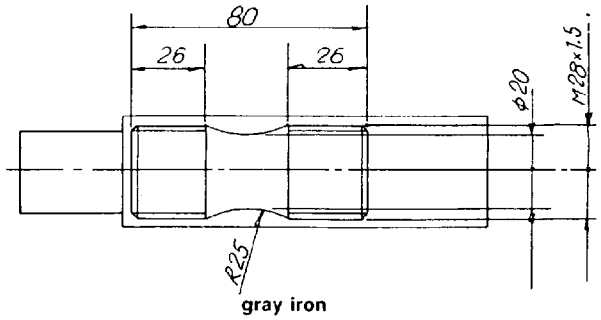
Figure 8 shows the dimensions of welded test bars of homogeneous gray, malleable and ductile iron. If the metals welded together were dissimilar, test bars with a 20 mm diameter, useful length of 200 mm and total length of 250 mm were used. Elongation of these bars was measured over a length equal to five times the diameter.

In all cases, care was taken to insure that the weld zone was located evenly with respect to the median line of the smallest diameter, in the case of gray iron, and in the useful length of the bar for the other types of cast iron or other mixed metals.

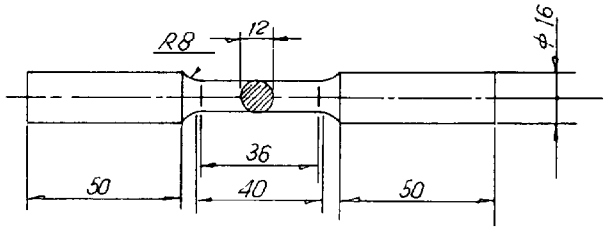
For tests of impact strength, notched and unnotched test bars were produced in such a way that for the notched bars the notch is located in the weld zone. For this test, a pendulum quite a bit smaller than the usual charpy hammerhead was used and the values are indicated specially in kg cm^2 instead of m-kg cm^2, these latter units being used only in the case of particularly high results.

Hardness tests were conducted by either the rockwell or brinell method, according to the hardness of the metals tested, longitudinally up the test bar from one side to the other of the metals welded, starting from the zones of unaltered base metal and passing through the weld zone.

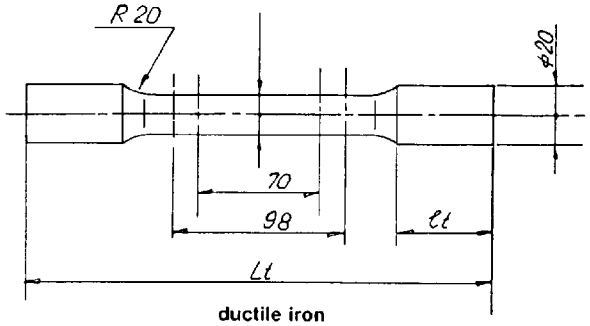
Metallographic examination was carried out on all weld sections starting from the base metal, which was not thermally influenced, and going across the weld zone to the other side consisting of the same or different material. A series of photomicrographs was taken of which, in this paper, only the most significant ones are shown.



gray iron



malleable iron



ductile iron

Fig. 8. Dimensions of welded test bars.

Results of Test Welds Between Homogeneous Cast Irons

Flake Graphite Gray Cast Iron Type G 20 UNI 5007/69 The type of cast iron tested had the following composition:

3.08% C	1.45% Si	0.73% Mn	0.065% P
0.016% S	0.019% Cr	0.27% Ni	0.10% Mo
0.10% Cu			

where Cr, Ni, Mo and Cu should be considered as impurities arising from tramp metals in the charge.

The tensile strength of the base iron is about 20 kg mm^2 and its hardness around 200 HB (HRB-10-3000-30 kg mm^2). The structure of the metal consists of type A graphite flakes in a pearlitic metal matrix, for the most part.

Several welded test bars, according to the procedure described above, were prepared; starting with these, we obtained 2 tensile strength test bars, 2 impact strength test bars, and 2 bars

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Table 1. Tensile Strength and Impact Strength of Flake Graphite Gray Cast Iron Welded Test Bars

Bar No. / Cooling Condition After Welding		1	2	3	4	5	6	7	8	9
		in air			beneath ashes			annealed		
Dimensions	d	19.8	19.9	—	19.9	19.9	—	20	19.9	—
	a	—	—	10	—	—	10	—	—	10
	b	—	—	8	—	—	8	—	—	8
Section in mm²		307.9	311	80	311	311	80	314	311	80
Useful Part, mm		UNI	UNI	40	UNI	UNI	40	UNI	UNI	40
Total strength, kg		5600	5650	—	4700	5900	—	6700	6650	—
Unit strength, kg/mm²		18.2	18.1	—	15.1	19	—	21.1	21.4	—
Impact strength, kg/cm²		—	—	17 18			19 20	—	—	25 25

for hardness tests and for microscopic examination of structure. As stated earlier, the series of test bars was subjected, after welding, to different cooling cycles already described. The mechanical properties obtained from the tensile and impact test bars are reported in Table 1.

Figure 9 shows a macrograph of the longitudinal section of a weld joint and the superimposed diagram represents the development of Rockwell B hardness at 1/16-100 kg. Development of the hardness in a similar fashion was found in all the test bars.

Upon examination of the experimental results, we can state that:

- the tensile and impact strength of the welded joints reached values very close to those of the base cast iron, which implies that there is no drop in strength of the metal following welding.
- the values of tensile and impact strength undergo a gradual improvement, quite clearly, on adopting a slower cooling rate after welding, still better upon performing an annealing heat treatment after the welding cycle. Actually, whereas the strength of welded test bars cooled in air (1 and 2), Table 1, is around 18 kg mm^{-2}, that of annealed test bars (7 and 8) reaches higher values of 21 kg mm^{-2} and it is also the same for impact strength which improves from 17-18 kg cm^{-2} to 25 kg/cm^{2} for the annealed test bars.
- the hardness values, which in the base metal are on the order of 83-88 HRB, increase gradually in the thermally influenced zone until they reach a maximum value of 108-112 HRB in the contact zone of the junction.

The thermally influenced zone, shown in the macrograph, has a limited extension, and the differences in hardness values along the length of the test bar are extremely small, so that it is thought that the formation of very hard elements does not take place. The differences in hardness between the different test bars, as distinguished by the cooling conditions, are very modest, even if there exists, in the weld zone, a hardness slightly greater in the test bars cooled only in air, as compared with those immediately subjected to an annealing heat treatment.

Figure 10 (a and b) shows the structure of the weld joint of test bar No. 1, unetched and with Nital etch, at 200X. From these photomicrographs we can state that in the junction and contact zone between the two welded surfaces the segregations of flake graphite have entirely disappeared, being squeezed out during the flashing, and there only remain globules of graphite in a nearly spheroidal form.

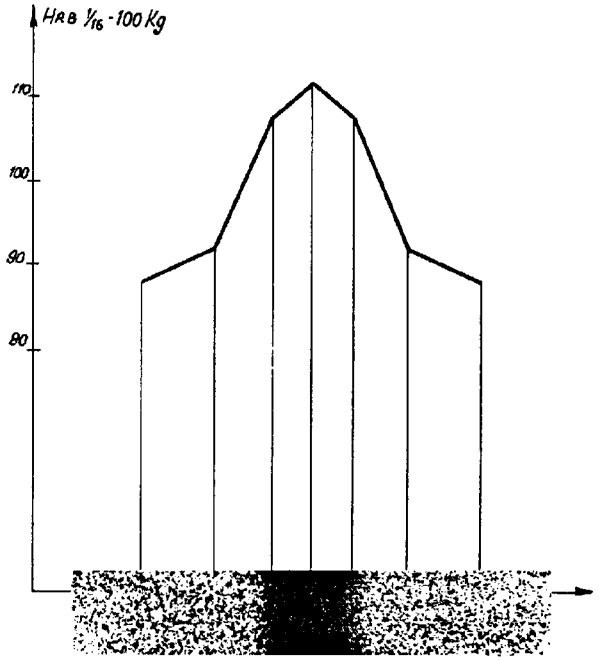


Fig. 9. Hardness values for a weld between homogeneous cast irons (G 20 UNI 15007/69).

The weld junction and the atomic interpenetration between the two joined surfaces prove to be complete and such as to insure a good and uniform continuity of metal.

The structure of the metallic matrix in the contact zone and the weld junction presents a clear network of cementite, which follows the shape of the prior austenite grains. The phenomenon is explained clearly upon studying the sketch of Fig. 3. Actually, following the high temperature, which is practically that of the molten metal, reached at that point, the free carbon or graphite which has not been expelled during the flash welding is almost entirely dissolved in the austenite by diffusion in the solid and semi-liquid states at temperatures between 900 and 1200C (1652 and 2192F). Following rapid cooling, there is no time for it to isolate as a graphite shape and it remains in the combined state as cementite, which partially segregates at the margins of the austenite crystals, while the remaining combined carbon gives rise to a matrix that is entirely and finely sorbitic, practically free from segregations of free graphite.

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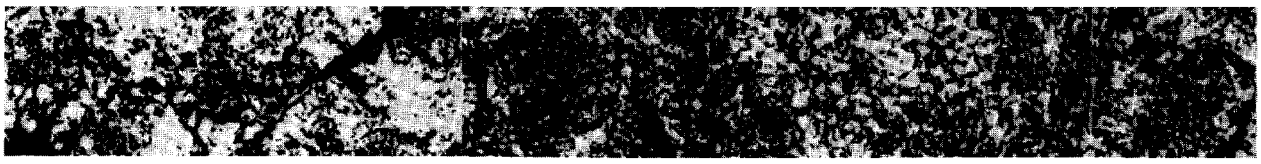


Fig. 11. Structure of the weld zone between two samples of gray iron G 20 UNI 15007/69 after annealing treatment on the welding machine (200X). Nital etch.

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Table 2. Tensile Strength and Impact Strength of GMN 35 Malleable Iron Welded Test Bars

Bar No.		1	2	3	4	5	6	7
Cooling Condition After Welding		in air		beneath ashes		annealed		
Dimensions mm	d	12	—	12	—	12	12	—
	a	—	10	—	10	—	—	10
	b	—	8	—	8	—	—	8
Section mm²		113	80	113	80	113	113	80
Useful part, MM		36	40	36	40	36	36	40
Total strength, kg		3850	—	3850	—	3350	3.600	—
Unit strength, kg/mm²		34	—	34	—	29.6	31.8	—
Elongation,% on 5 D		2.8	—	3.3	—	1.1	2.2	—
Impact strength,kg/cm²		—	22	—	22	—	—	80
Observations:		All bars broke outside the useful portion and outside of the weld						

Table 3. Tensile Strength and Impact Strength of GMN 45 Malleable Iron Welded Test Bars

Bar No.		1	2	3	4	5	6
Cooling Condition After Welding		in air		beneath ashes		annealed	
Dimension mm	d	12	—	12		12	—
	b	—	10	—		—	10
	a	—	8	—		—	8
Section mm²		113	80	113		113	80
Useful part, mm		36	40	36		36	40
Total strength, kg		4050	—	4550		4600	—
Unit strength, kg/mm²		35.8	—	40.2		40.6	—
Elongation, % on 5 D		1.6	—	2.2		2.5	—
Impact strength, kg/cm²		—	36	—	25	—	29
Observations		- = broken outside the useful part and outside the weld. 4,5,6: broke outside the weld					

Table 4. Tensile Strength and Impact Strength of GMN 65 Malleable Iron Welded Test Bars

Test Bar No.		1	2	3	4	5	6	7
Cooling Conditions After Welding		in air		beneath ashes		annealed		
Dimensions mm	d	12	—	12	—	12	12	—
	a	—	10	—	10	—	—	10
	b	—	8	—	8	—	—	8
Total section, mm		113	80	113	80	113	113	80
Useful part, mm		36	40	36	40	36	36	40
Total strength, kg		7000	—	6650	—	6400	5650	—
Unit strength, kg/mm²		62	—	58.8	—	56.5	49.8	—
Elongation,% on 5 D		2.2	—	3.0	—	2.5	1.7	—
Impact strength,kg/cm²		—	34	—	24	—	—	31
Observations:		broke outside the weld		broke outside the weld				

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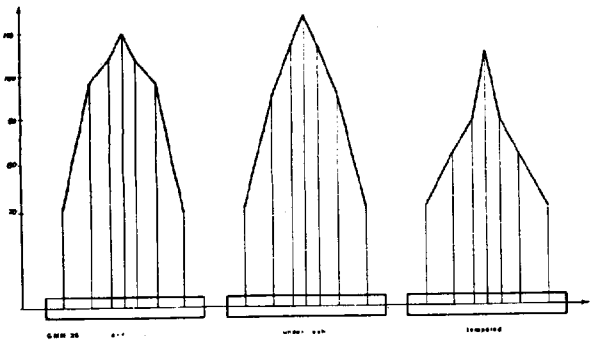


Fig. 12. Hardness values through longitudinal sections of test bars of GMN 35 iron, different cooling methods and heat treatment.

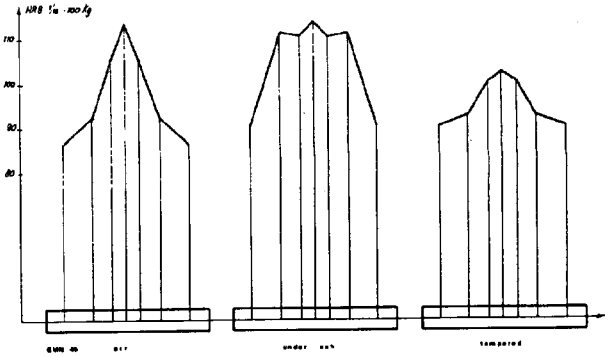


Fig. 13. Hardness values through longitudinal sections of test bars of GMN 45 iron, different cooling methods and heat treatment.

On moving away from the junction zone, the structure in the thermally influenced zone passes gradually from sorbitic to pearlitic in the furthest zones, where the base iron has not been subject to any thermal influence.

The photomicrograph of Fig. 11 shows the structure of the same zone shown in Fig. 10 of a test bar which has been submitted to annealing heat treatment on the welding machine, after which it was allowed to cool naturally. One may observe a clear beneficial influence, even though the treatment time is extremely limited, since the destruction of hardened structure from cementite formations observed in Fig. 10 can be seen, and the formation of a very fine uniform pearlite structure, to which are linked improved mechanical properties.

Results of Test Welds of Malleable Irons GMN UNI 3779/69 — The blackheart malleable irons listed in the UNI 3779/69 table are characterized respectively by a metallic matrix structure consisting of ferrite, pearlite and ferrite, and globular pearlite, which correspond to the minimum breaking strengths of 35-45 and 65 kg/mm^2; minimum elastic limit strengths of 22-26-39 kg/mm^2 and minimum elongation of 14-7-3% on 5 diameters and brinell hardness of 120/160-150/210-210/250 HRB for the three types of cast iron which we have subjected to testing, namely GMN 35, GMN 45 and GMN 65.

In these cases, the welding experiments were carried out on 25 mm diameter cylindrical test specimens generated by turning the parts as indicated in Fig. 5.

The smallest diameter of the weld specimens, in relation to those of gray iron, affects the conditions under which the weld is made, especially the time and the level of heating, the surface area of the zone that is thermally influenced and subject to structural transformations, and the speed of cooling.

Even for these types of cast iron, some test bars were allowed to cool naturally in the air after welding, while others were cooled in a bed of ashes, and still others were subjected to an annealing heat treatment as described previously.

Tables 2, 3, and 4 show the mechanical properties obtained from tensile and impact strength test bars, and Fig. 12, 13, and 14 show the hardness diagrams along the length of the test bars, from the three types of cast iron tested; GMN 35, GMN 45 and GMN 65.

Note that in almost every case the rupture point is outside of the head area, that is, completely within the base metal, which demonstrates that the welded zones and the zones thermally influenced by welding have a higher strength than that of the respective base cast irons. This is why the strength values agree with those that are typical for each type of cast iron.

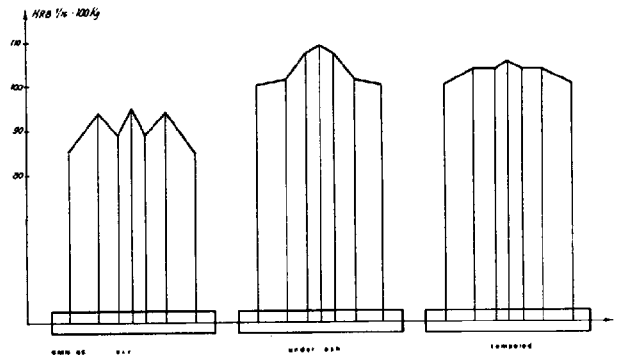


Fig. 14. Hardness values through longitudinal sections of test bars of GMN 65 iron, different cooling methods and heat treatment.

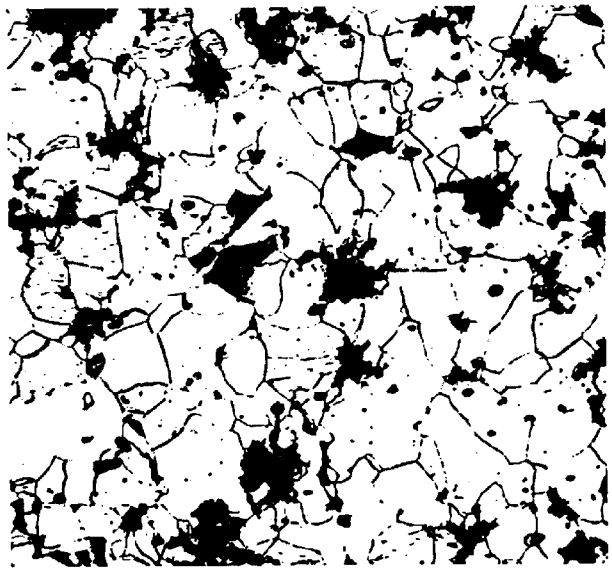


Fig. 15. Structure of cast iron GMN 35 in thermally unaffected zone (200X). Nital etch.

The elongation measurements are significantly lower than the minimums prescribed by the UNI table. The reduction of these values is also due to the fact that the weld zone and the thermally influenced zone, which are stronger than the base metal, cannot be deformed.

The thermally influenced zones are found to be considerably more extensive than those observed in homogeneous gray cast

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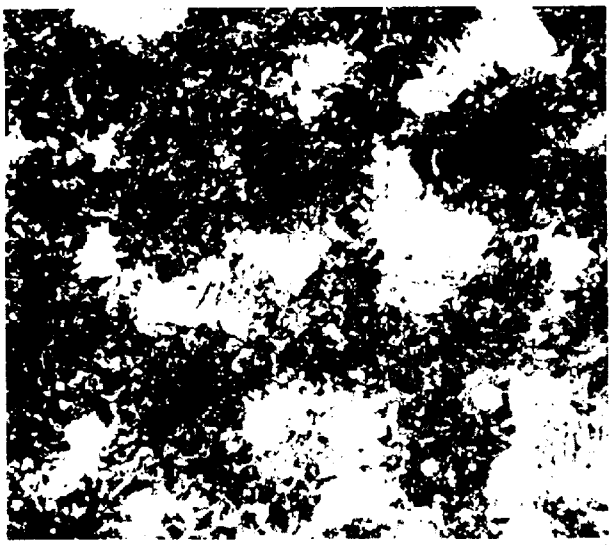


Fig. 16. Structure of cast iron GMN 35 in thermally influenced zone (200X). Nital etch.

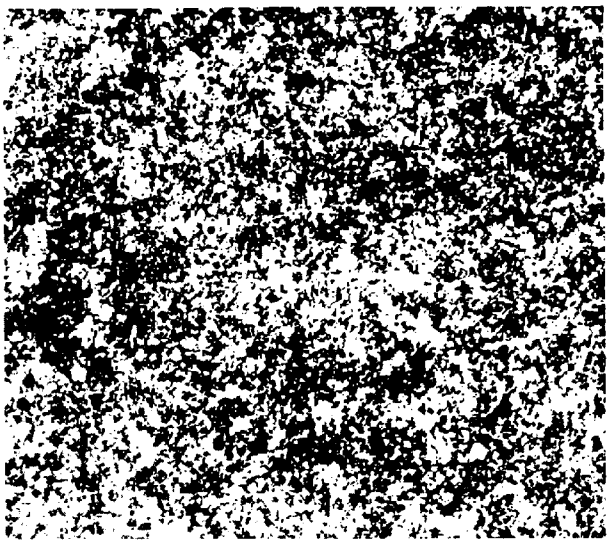


Fig. 18. Structure of cast iron GMN 65 in thermally unaffected zone (200X). Nital etch.

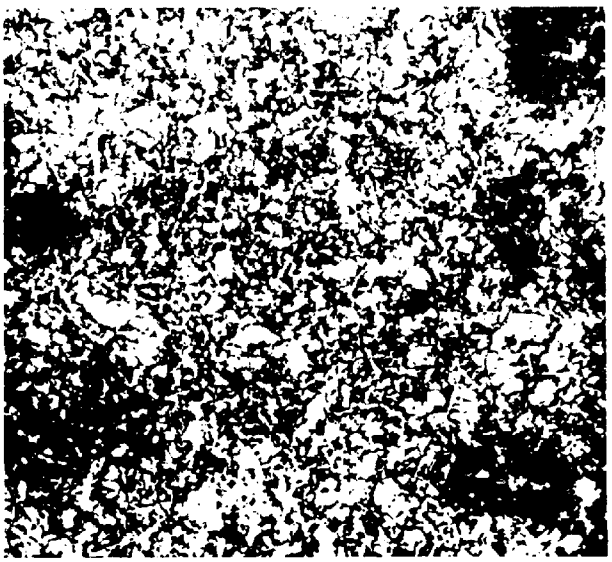


Fig. 17. Structure of cast iron GMN 35 in zone of maximum temperature (200X). Nital etch.

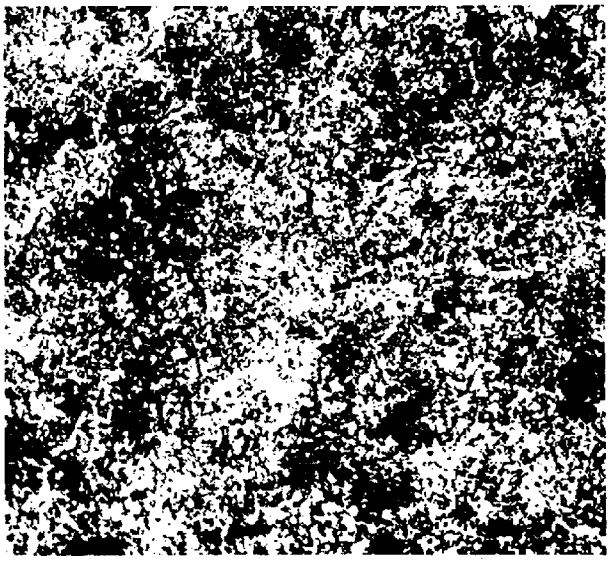


Fig. 19. Structure of cast iron GMN 65 in weld zone (200X). Nital etch.



Fig. 20. Longitudinal structure of welded test bar, GMN 65 cast iron (200X). Nital etch.

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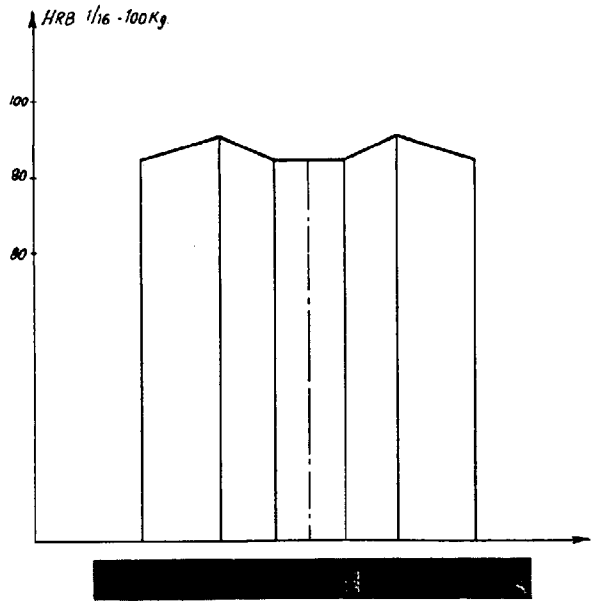


Fig. 21. Hardness through longitudinal section of test bar, GMN 65 cast iron.

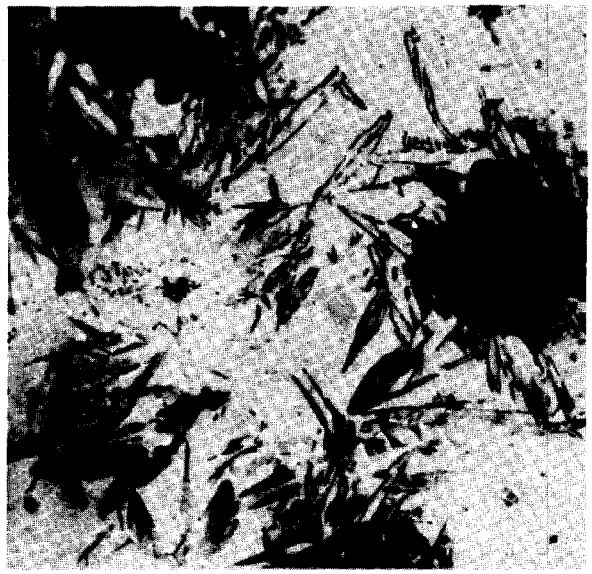


Fig. 22. Structure of corrosive resistant ductile iron (850X). Nital etch.

iron. This can be explained by the smaller diameter of the test specimens subjected to welding, but especially due to the fact that these types of cast iron, because of their typical structure, are characterized by a thermal conductivity significantly higher than that of gray iron with flake graphite, which results in an extension by conduction of the heated surface in a given unit of time. For this reason, it was not possible to make continuous photomicrographs for the entire length of the structurally transformed zone. On the other hand, metallographic examination has established that the zone of the junction and those zones thermally affected by the welding show, for all types of malleable cast iron, a practically identical structure, which is equally confirmed by the measurement of hardness values from the various test bars. This is why we think that it is not necessary to describe all of the types of test bars tested. We are taking care, however, to illustrate the most significant structures of the test bars of two extreme types of malleable cast iron, that is, GMN 35 and GMN 65. The photomicrograph of Fig. 15 shows the structure of cast iron GMN 35 in the unaffected zone, etched in Nital, at 200X, consisting of temper carbon, flakes, and nodules in an entirely ferritic matrix.

Figure 16 shows the structure of the same test bar welded in the thermally influenced zone (at 200X and etched in Nital). Beyond the characteristics produced by the welding cycle, the graphite flakes (nodules) are partially taken into solution in the γ iron, and in the following phase of cooling, a troostitic-pearlitic structure forms in the zones near the residual graphite flakes.

The phenomenon of graphitic carbon going into solution and the formation of troostite-sorbite structures (stronger than that of the original base metal) extends and gradually becomes complete upon approaching the contact zone for the weld junction, where the maximum temperatures are reached and where (Fig. 17 - 200X, Nital etch) the disappearance of graphite flakes is complete and the matrix structure is entirely troostitic-sorbitic, with traces of a dispersed cementite network which follows the edges of the prior austenite grains.

Figure 18 shows at 200X, Nital etch, the structure of GMN 65 cast iron in the zone which is not thermally affected. This displays graphite flakes in a metallic matrix consisting of typical fine pearlite in the quenched and tempered state.

Figure 19 shows the structure of this same cast iron (at 200X, Nital etch) in the junction zone, which is found to be practically destitute of graphite flakes and which consists of troostite and of pearlite with an improved appearance in every way analogous to those of the thermally affected zone and very similar to those of the base cast iron. Microscopic examination of the different test bars allows us to state that there are no substantial differences between the welded test bars which have been cooled in air and test bars cooled in ashes and annealed.

Results of Test Welds of Ductile Iron Type GS 55/5 UNI 4554 — The ductile iron used for our tests has the following composition: 3.43% C, 1.65% Si, 0.45% Mn, 0.059% P, 0.013% S, 0.66% Ni and 0.082% Mg and the following mechanical properties:

Rupture strength = 43.6 kg/mm^2
Elongation on 5 diameters = 5%
Brinell hardness, HRB 10/3000/30 = 235.

The metal structure is comprised of well-shaped nodular graphite, of the type OA UNI 3775, well-distributed in a matrix consisting of wide ferrite zones mixed with those of pearlite.

In this case, having established that there is no substantial difference between test bars whether cooled in air, in ashes, or annealed, we confine ourselves to reporting the results of the test bars which were welded and annealed.

Figure 20 shows the structure of one of these test bars starting from the non-affected zone and progressing for the entire weld zone, magnified 200X and prepared with a Nital etch. We can see that the junction and the interpenetration of welded surfaces are perfect. In contrast with graphite flakes, the spheroids of graphite are equally present in the weld zone but with their dimensions somewhat reduced. Nonetheless, here also they are partially dissolved in the iron during the hottest phase of the weld cycle which causes carbon enrichment of the surrounding metallic matrix. As a consequence, in the following phase of cooling, this favors formation of a very fine pearlitic structure, where the following anneal brings about elimination of the few free carbides which were formed during the rapid cooling phase.

Figure 21 shows the photomicrograph and the corresponding diagram of hardness along the length of the section. In contrast with that observed for other types of cast iron, here there exists

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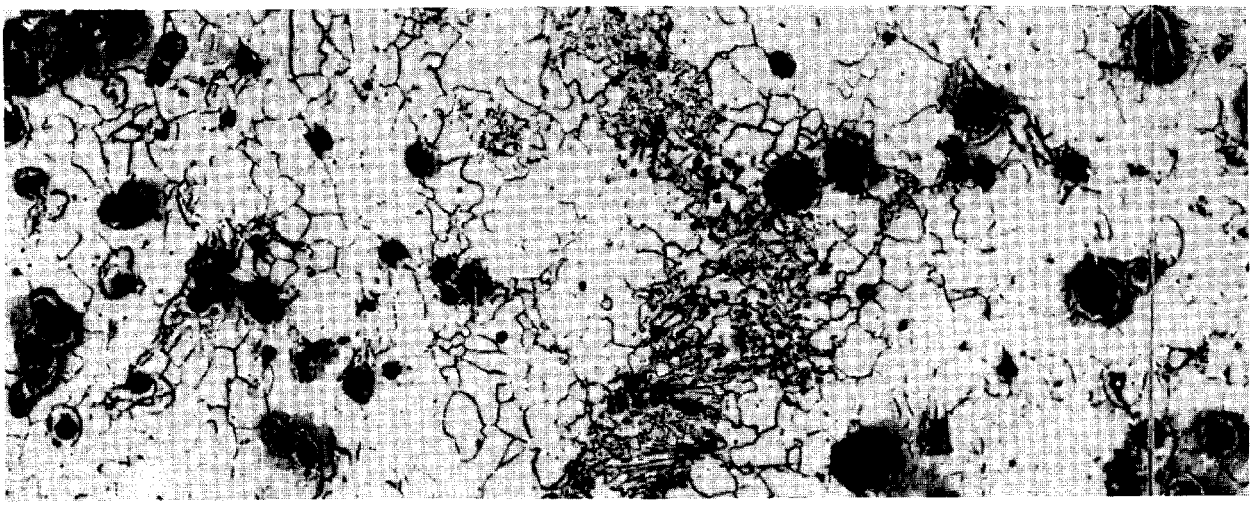


Fig. 23. Structure of weld along its length (200X). Nital etch.

an almost complete uniformity of hardness for the length of the entire test bar.

In Table 5 the mechanical properties of test bars obtained from the welded test specimens are reported.

Concerning these results, we may conclude that the weld junction is perfectly successful and illustrates good mechanical and structural properties.

Results of Test Welds of Corrosion-Resistant Alloyed Ductile Iron – The experiments were extended with a type of ductile iron alloyed with Ni-Cr-Si-Cu and resistant to specific corrosive environments.

The composition of this iron is as follows: 2.72% C, 4.22% Si, 0.35% Mn, 0.98% Cr 14.62% Ni and 0.76% Cu.

The structure of the cast iron, revealed by etching and at 850X, shown in Fig. 22, consists of well-shaped graphite spheroids surrounded by large martensite needles in an austenitite matrix.

Figure 23 shows the structure of the welded joint along all its length starting with the unaltered base cast iron (at 200X, Nital etch). Upon observing it, we can state that:

- In the thermally affected zone, the needles of martensite disappear little by little, being redissolved during the hot phase of the weld cycle.
- The graphite spheroids remain unaffected in all the thermally influenced zone and even in the weld zone, even though their dimensions are reduced with respect to the original dimensions, which confirms that this form of more compact graphite resists solution in the austenite better than the flake and floccular shapes.
- The line of the weld junction, where the melting temperature has been reached, has been delineated by a continuous line of carbides, which are isolated during the cooling and whose formation was probably favored by the presence of Cr in the alloy, Cr being a notorious carburizing element.
- In the thermally affected zone immediately next to the weld and the line of carbides, the structure of the metallic matrix appears to have taken on a clearly ferritic aspect. The formation of ferrite by elimination of carbon in the form of carbides and the redissolving of martensite needles could have been favored by the high Si content, an element having a powerful ferritizing action, and by Cr, an element which is carburizing and at the same time ferritizing.
- It is evident that the presence of carbides and ferrite right beside the austenitic and martensitic structures constitutes a distinct collection of structures which gives rise to a deterioration of the corrosion resistance properties and, as a consequence, it would be necessary to carry out a follow-up experiment to find a suitable heat treatment after welding which is more suited to reconstituting a uniform and a homogeneous structure.

Table 5. Mechanical Properties of Welded Test Bars of Ductile Iron

Cooling Conditions After Welding	Annealed
Dimensions d a b	15 mm
Section mm²	176,7 mm²
Useful part, mm	75 mm
Overall strength, kg	7000 kg
Unit strength, kg/mm²	39,6 kg/mm²
Elongation, %	0,5%
Impact strength	60 kg/cm

Remarks: It was not possible to extend the experiments with this type of cast iron advantageously, for lack of material.

The photomicrograph of Fig. 24 and the joint diagram of hardness values along the length of the welded test bar confirm perfectly what has been observed for the structures and permits us to make the following observations:

- The extension of the thermally affected zone is, in contrast to that which has been noted for flake graphite cast irons and, especially, for the malleable cast irons, extremely limited, practically reduced to a thin line with respect to the contact surfaces of the weld joint. This can be explained by noting that the thermal conductivity of austenite, especially if this latter is alloyed with Ni and with Cr, is

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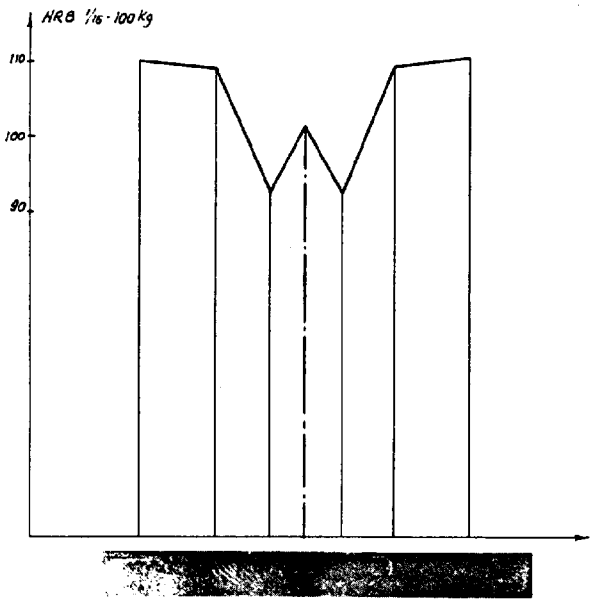


Fig. 24. Hardness through longitudinal section of test bar. Gray iron on left, ductile iron on right.

Table 6. Mechanical Properties of Welded Test Bars of Corrosion-Resistant Alloyed Ductile Iron

Cooling Conditions After Welding		Annealed
Dimensions d	12	
b	—	10
a	—	8
Section mm²	113	80
Useful part, mm	36	40
Overall strength, kg	1950	
Unit strength, kg/mm²	17.2	
Elongation, %	—	> 110
Observations:	With defects	Not Broken

considerably less than that of the other experimental structures.

- In contrast to what has been observed in the weld joints of other types of cast iron, the maximum hardness values are found in the base cast iron, that is, corresponding to the austenite-martensite structure. At the weld junction line, that is, in the zone of carbide segregation, the hardness is lower and the minimum hardness is found in the intermediate thermally affected zones, where we have observed the presence of ferrite.

Table 6 shows the mechanical properties obtained from test bars produced from the welded test specimens of this type of cast iron recalling, however, that in the case of tensile strength test bars with many defects, it has not been possible to determine the exact value of the breaking strength. If some defect had not been present, this value would have been higher, as predicted by the extremely high impact strength value where the test bar bent without breaking.

Results of Test Welds Between Heterogeneous Ferrous Metals

Flash welding of ferrous metals of different natures, and by this is understood the different qualities of cast iron and of steel, should give rise to efficient weld junctions characterized by good strength, comparable to those of the base metals. Actually, the characteristics of the metals can influence the welding technique and the final characteristics of the weld joint very little.

The melting points and the corresponding temperatures of softening of the different cast irons are the same, so that the compression of parts requires similar pressures. On the other hand, in the case of steel, the squeeze temperature is higher than that of cast iron, so that it must have a different position in the jaws of the welding machine as shown in Fig. 6.

The electrical and thermal conductivity of the different types of cast iron present few differences, but it is necessary to take this into account when beginning the flash phase and determining the amount of compression to use. The noticeable difference in elongation between gray cast iron with flake graphite and the malleable cast irons, particularly ductile iron, and even more steel, can be the cause of fractures at the weld joint unless appropriate measures are taken. The free carbon diffusion phenomena, most especially in the case of welding cast irons with steel, can significantly modify the properties of the weld joint.

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Table 7. Tensile Strength and Impact Strength of Test Bars of Gray Iron Welded to Various Ferrous Metals

		1a) Ductile Iron		1b) Malleable Iron		1c) Steel	
		Tensile	Impact	Tensile	Impact	Tensile	Impact
Dimensions	d	15	—	15	—	15	—
	a	—	10	—	10	—	10
	b	—	10	—	10	—	8
Section mm²		176,6	100	176,7	100	176,5	80
Useful part, mm		75	40	75	40	75	40
Total strength, kg		3750	—	3100	—	3850	—
Unit strength, kg/mm²		21,2	—	17,5	—	21,8	—
Elongation, % on 5 D			—	—	—	—	—
Impact strength, kg/cm2			27-27	—	24-27	—	23-23,8
Observations:		All specimens broke beneath the head on the cast gray iron side					

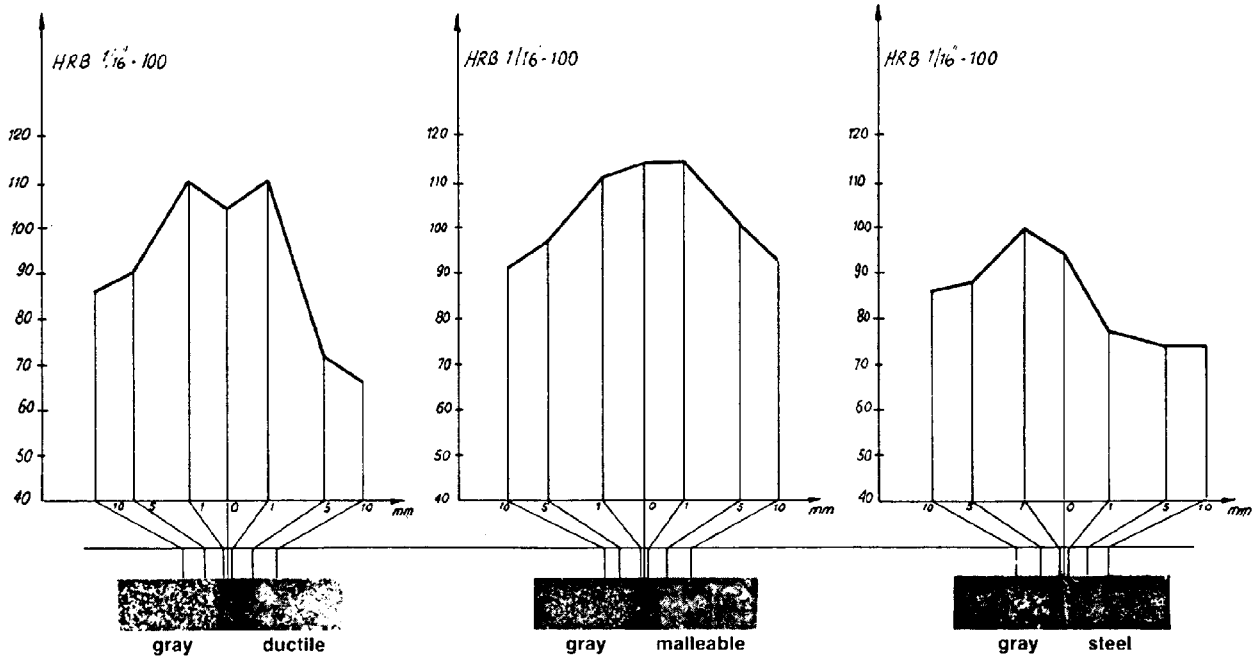


Fig. 25. Hardness through longitudinal sections of test bars.

Results of Mechanical Tests

Table 7 shows the results obtained in the tensile and impact strength tests on welded test bars made of gray iron and specimens of other ferrous metals.

We observe that in all the tests the welded junction zone resisted the tensile stress perfectly and that the rupture took place beneath the head of the test bars on the gray iron side, at a breaking strength equal to that of the metal.

The welded joints do not show the slightest trace of cracking and therefore they demonstrate a strength notably superior to that of the weaker of the two metals joined.

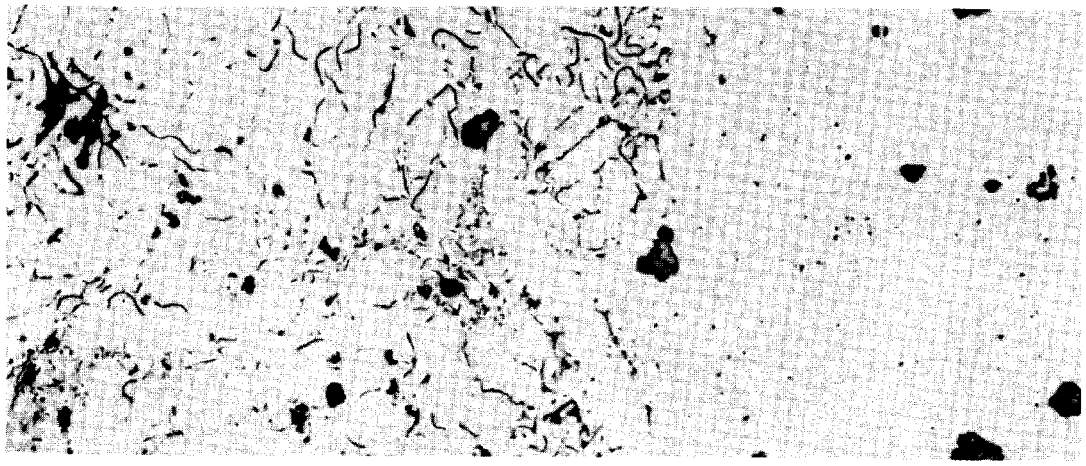
The test bars show no elongation, even in the case of welding with steel, because they break in the gray iron part at a breaking strength below the elastic limit of the other metal, which has not even begun to deform.

The rupture of impact strength test bars, some of them notched and others unnotched, always took place in the weld section, and the values obtained are very significant.

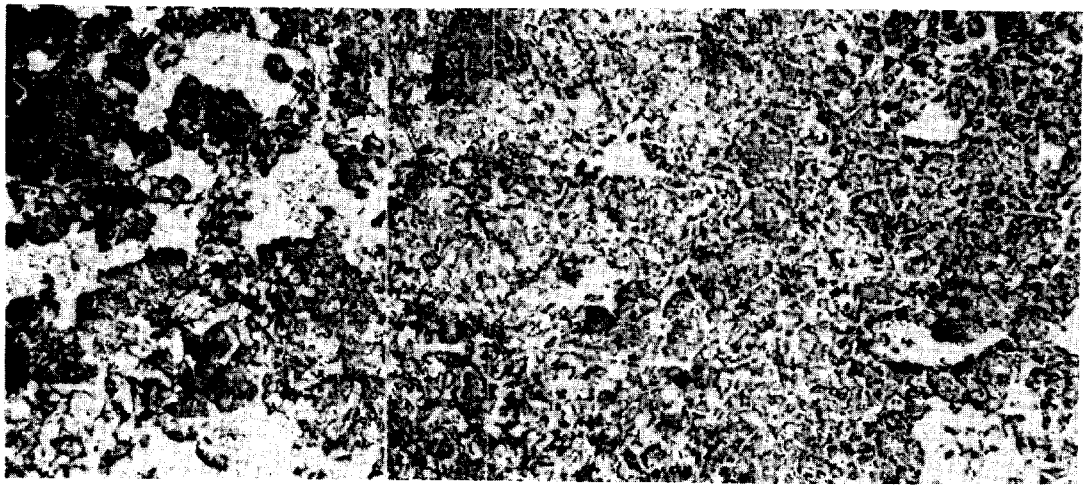
Figure 25 shows the photomicrographs of longitudinal sections of welded joints and the superimposed diagrams graphically represent the rockwell HRB hardness 1/16 100, starting from the unaffected base metal and crossing the weld joint. From these diagrams, it turns out that the hardness of the welded zone is always the highest, which proves that it has the greater resistance to breaking. On both sides, the hardnesses diminish until they are the same at a distance of around 10 mm from the weld joint, as the typical hardness values of the base metal.

The photomicrographs of Fig. 26 (a and b) show the structure in the joints of the welded test bar between the flake graphite cast iron and the ductile iron, at a magnification of 200X.

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(a) Unetched.



(b) Nital etch.

gray iron ductile iron

Fig. 26. Structure in the weld zone between flake graphite cast iron and ductile iron (200X).

Structural observation completely proves out what we have already seen previously in the case of welding various types of cast irons with themselves, that is:

- In the zone of the joint, the free graphite segregations have practically disappeared, especially those of flake graphite squeezed out by the electric arcs during flashing, and partially dissolved in the austenite just below the melting temperature. The junction and the atomic interpenetration of the surfaces of two different kinds of cast iron appears to be extremely close, such as to insure the most complete and uniform continuity between the metals joined.
- During the following phase of rapid cooling, in the joint zone, a structure forms which is characterized by a cementite network distributed along the edges of prior austenite grains, enclosed within a matrix structure of sorbite without any trace of free graphitic carbon.
- On moving away from the joint zone on both sides, in the thermally affected zone of both types of cast iron, the solution of flake graphite, to the left, and of spheroidal graphite, at the right, appears to gradually decrease. The carbon thus dissolved, because of the very rapid cooling which follows, does not have time to separate from the austenite in the form of cementite, but remains trapped in the alpha iron and gives rise to localized martensitic shapes.
- A final annealing treatment, carried out directly on the welding machine, is a means of redissolving all of the hard structures and of transforming them into a uniform, finely pearlitic structure to which better toughness properties correspond.

The photomicrographs of Fig. 27 (a and b) show the structure in the junction zone between flake graphite cast iron and malleable iron with floccular graphite at a magnification of 200X, unetched and with Nital etch. We note that:

- In the junction zone the free graphite segregations have almost disappeared and their number and their dimensions are greatly reduced, according to the phenomena already described. Nevertheless, the presence of a number of graphite flakes may be noted, tending to indicate that this graphite form is more stable and more resistant to combustion phenomena during the phase of flashing, of dissolving and of diffusion.
- The junction and the interpenetration between the different metals are uniform and such as to insure the continuity between them, thus justifying the good mechanical properties of the joint.
- In the same zone of welding and contact we find sparse formations of free carbides surrounded by a uniformly pearlitic structure especially on the flake graphite cast iron side.

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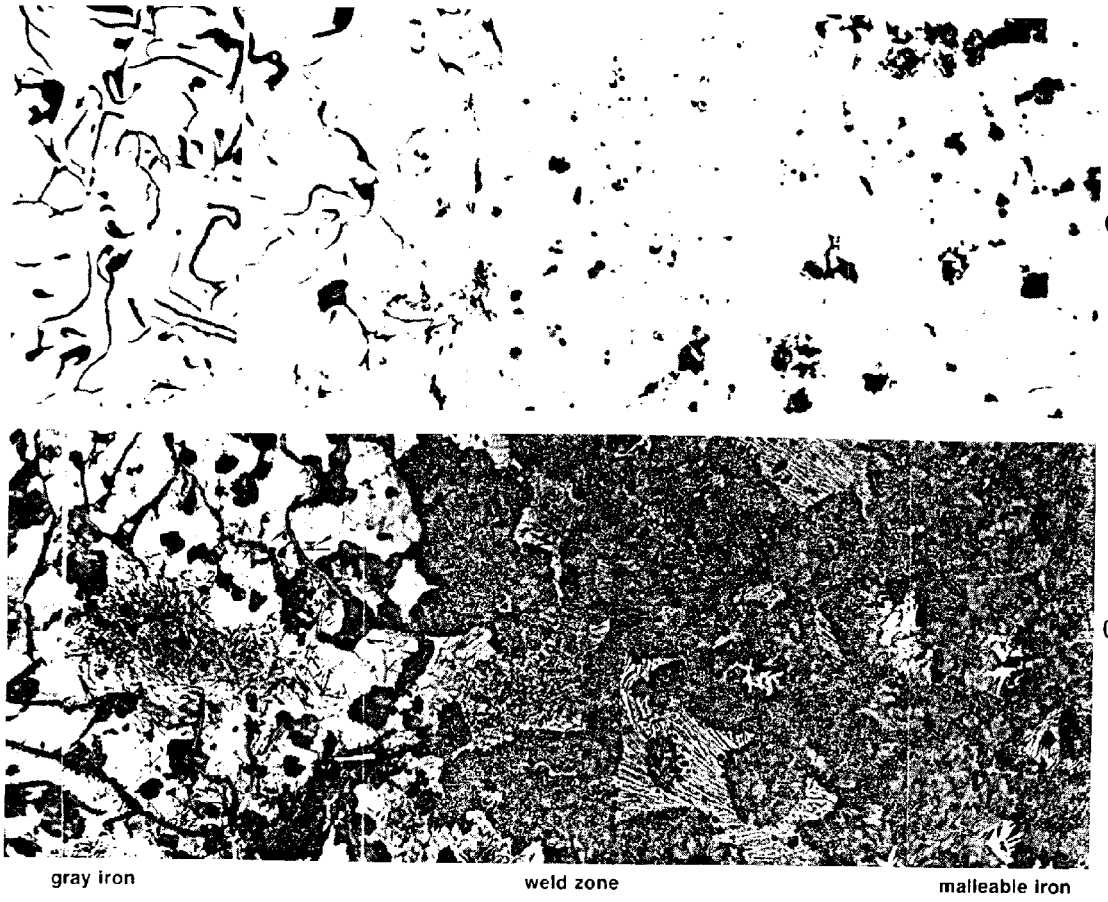


Fig. 27. Structure in the weld zone between flake graphite cast iron and malleable iron with floccular graphite (200X).

- By means of a final annealing treatment, these carbide formations are dissolved in a troostite structure.
- In the thermally affected zone of the flake graphite gray cast iron, we note a ferritic-pearlitic structure with good distribution of type A graphite flakes.
- On the malleable iron side, the structure is entirely pearlitic in the zones close to the joint, where the graphite flakes have undergone anew the phenomenon of dissolving in the metallic matrix. The quantity of pearlite decreases as we move away from the joint zone, this pearlite being always grouped around the graphite flakes (nodules) while the quantity of ferrite increases and the graphite flakes take on a better and less altered shape.

Figure 28 (a and b) show the structure of the joint between flake graphite gray cast iron and soft steel, unetched and with Nital etch, enlarged to 200X.

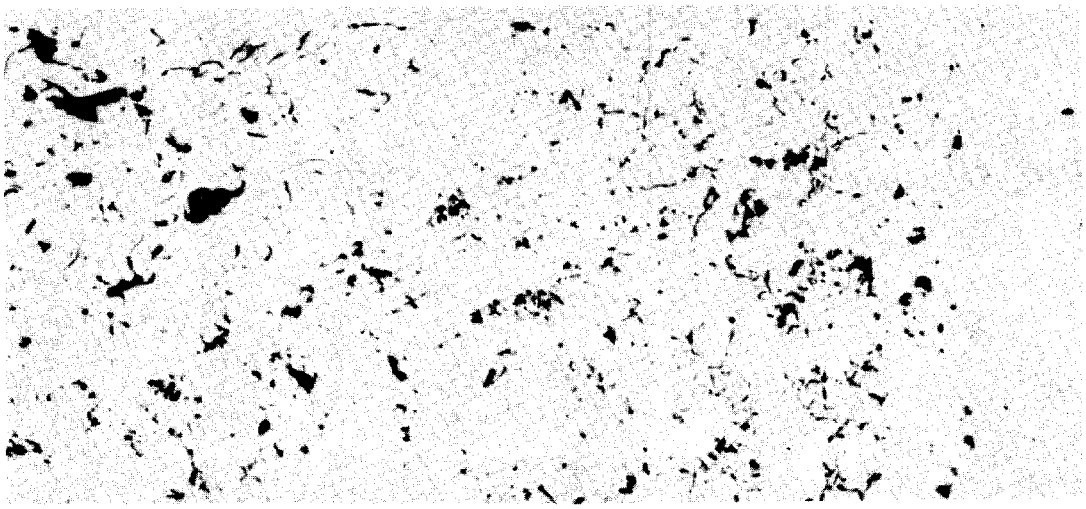
- The number of graphite flakes in the cast iron diminishes in the zone immediately next to the contact surface with steel, and the flakes take on a finely globular shape.
- A sharp separation is noted between the two metals, which are more clearly distinguished from each other than in the cases previously described.
- Nevertheless, the adherence between the two metals is complete and the interpenetration of the contact surfaces and of the junction at the atomic level is also complete, which justifies the strength properties, which do not find their limit except at the tensile strength level of the weakest metal, that is, of the gray iron.
- The structure on the cast iron side is entirely pearlite-troostite, so that this test bar had to be subjected to an annealing heat treatment after welding.
- The structure on the steel side is entirely ferritic, in keeping with the nature of steel with a low carbon content.

Table 8 gives the values obtained on the tensile and impact strength test bars produced from welded test specimens between cast iron with ductile and malleable iron and soft steel, and between that of malleable cast iron and steel.

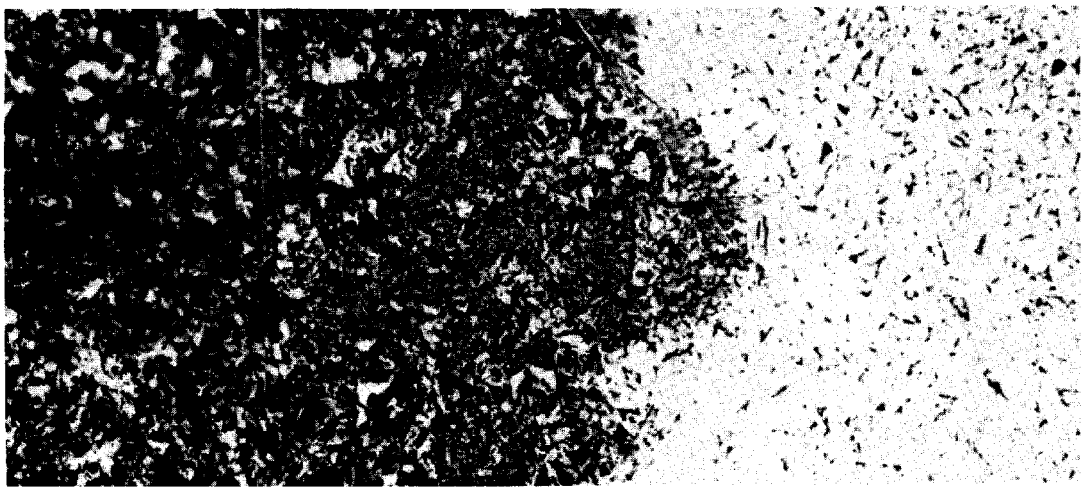
Figure 29 shows the macrograph of the longitudinal joint sections welded. The superimposed diagrams graphically represent the rockwell HRB hardness, I 16 100, along the length of the joint, starting with the two types of unaltered cast iron, in the three types of tests described above.

From an examination of the results, we can conclude that rupture of the tensile strength test bar, flash butt welded between a test specimen of ductile iron and a test specimen of malleable iron, took place immediately below the head of the test bars, outside of the useful zone, on the malleable iron side, at a load of 32.5 kg mm^2, equal to that of malleable iron type GMN 35, which proves that the joint withstood a load greater than the rupture strength of the weaker metal, that is, the malleable iron. Rupture took place at an elongation of 2% on the malleable iron

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(a) Unetched.



(b) Nital etch.

gray iron

steel
weld zone

Fig. 28. Structure of the weld between flake graphite gray iron and soft steel (200X).

side and had passed the elastic limit before breaking, as compared with the ductile iron which, having a strength and an elastic limit clearly greater than those of the malleable iron, did not exhibit any deformation. The impact test bar broke under a load of 1.2 m-kg cm^2 (not kilograms), which equally confirms the great toughness of the weld junction between the two types of cast iron.

The photomicrographs, Fig. 30 (a and b), unetched and with Nital etch, enlarged 200X, show the structure of the joint between the two types of cast iron. Contrast this with test bars showing joints where flake graphite gray cast iron is present. In this case, the presence of free graphite is observed for the entire length of the section examined, only its dimensions are smaller and its form is intermediate, between the nodular and floccular (flake) shapes. This confirms the observation already made that these two types of graphite support the thermal stresses of the weld cycle better than graphite of the flake type.

Also, the continuity and interpenetration between the two metals appears perfectly uniform and homogeneous. The structure of the metallic matrix at the joint is entirely troostite-pearlite. It changes gradually as we move away from the weld zone, until it takes on the original structure of the two types of cast iron at a distance of 10 to 15 mm from the weld line.

The test bar welded between ductile iron and steel broke under a load of 45.3 kg/mm^2, with an elongation of 4.5% in the ductile iron and 9% in the steel. The rupture took place at a distance of 2 or 3 mm from the contact section and the weld on the ductile iron side, that is, in the thermally affected zone, which is the weakest because of the presence of hard, fragile structural formations. The elongations may be considered typical for the two metals joined.

The impact strength values are also very high. All of this is evidence of the excellent properties of the welded joint.

The photomicrographs of Fig. 31 (a and b) show the structure in the junction zone between ductile iron and steel, unetched and with Nital etch, enlarged 200X.

In this case also, the bonding and the complete interpenetration of the two welded surfaces is excellent, with the most outstanding phenomenon being that of carbon diffusion from the more carburized metal (GS) to the less carburized one (steel).

At the contact surface on the cast iron side, the graphite nodules have disappeared and a dark zone is formed consisting of a finely divided martensitic and pearlitic structure, which is diffused into the ferritic matrix of the steel.

On the hardness diagram of Fig. 29, we see that the maximum hardness is reached along the whole length of the contact surface and the weld of the metals. The hardnesses drop off progressively on both sides of this surface until they reach the minimum

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Table 8. Tensile Strength and Impact Strength of Test Bars of Various Heterogeneous Ferrous Metals

		Ductile Iron Welded With: 2a) Malleable Iron		2b) Steel		Malleable Iron With: 3a) Steel	
		Tensile	Impact	Tensile	Impact	Tensile	Impact
Dimensions	d	15	—	15	—	15	—
	b	—	10	—	10	—	10
	a	—	10	—	8	—	8
Section mm²		176,7	100	176,7	80	176,5	80
Useful part, mm		75	40	75	40	75	40
Total strength, kg		5850	—	8000	—	5900	—
Unit strength, kg/mm²		32,5	—	45,5	—	35,6	—
Elongation, % on 5 D		2 GMN	—	4,5 GS 8,9 steel	—	5 GMN 1,8 steel	—
Impact strength, kg/cm²		—	1,2	—	8-9,5	—	1,10-1,37
Observations		Broke under the head on GMN side		Broke at 2-3 mm from the weld on Ductile Iron side		Broke at 8,5 mm from the point of the useful part on the GMN side	

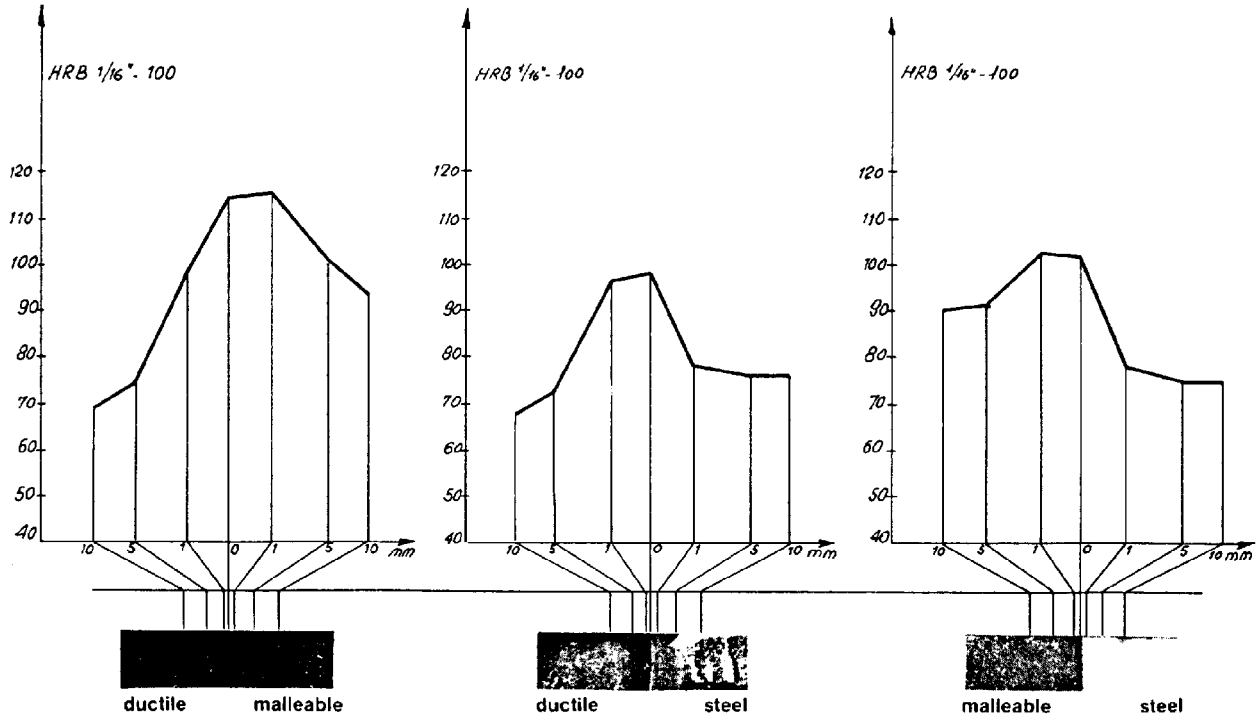


Fig. 29. Longitudinal sections showing rockwell hardness.

values corresponding to the typical values of the respective untransformed base metals.

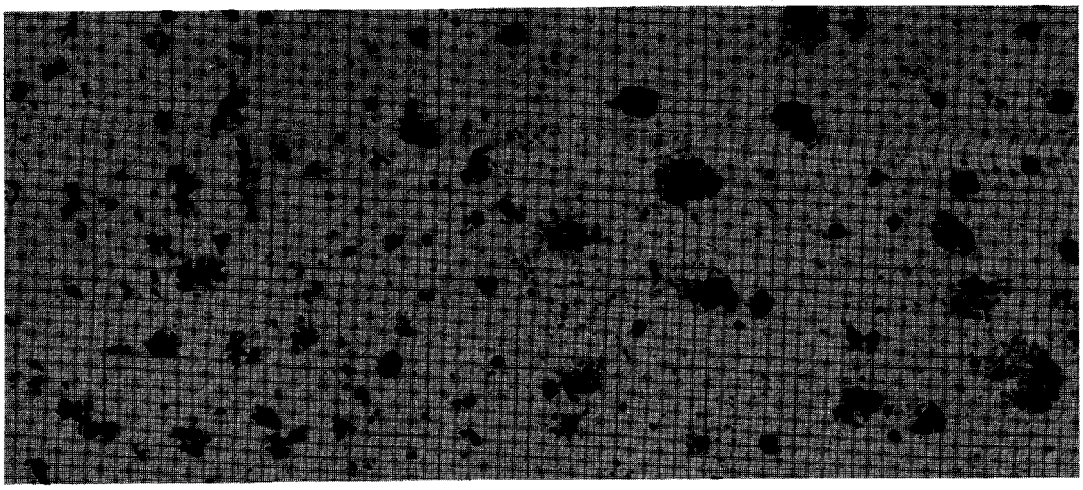
The test bar welded from malleable cast iron and steel broke at its midpoint in the malleable iron section, at a load of 35.5 kg mm², which is normal for the cast iron GMN 35 used in this research. Elongation was 5% in the malleable iron part and 1.8% in the steel portion, so that the load applied to the steel at the moment immediately after rupture was not sufficiently high to cause a significant amount of deformation.

The impact strength values of 1.1 and 1.4 m-kg cm² are very high.

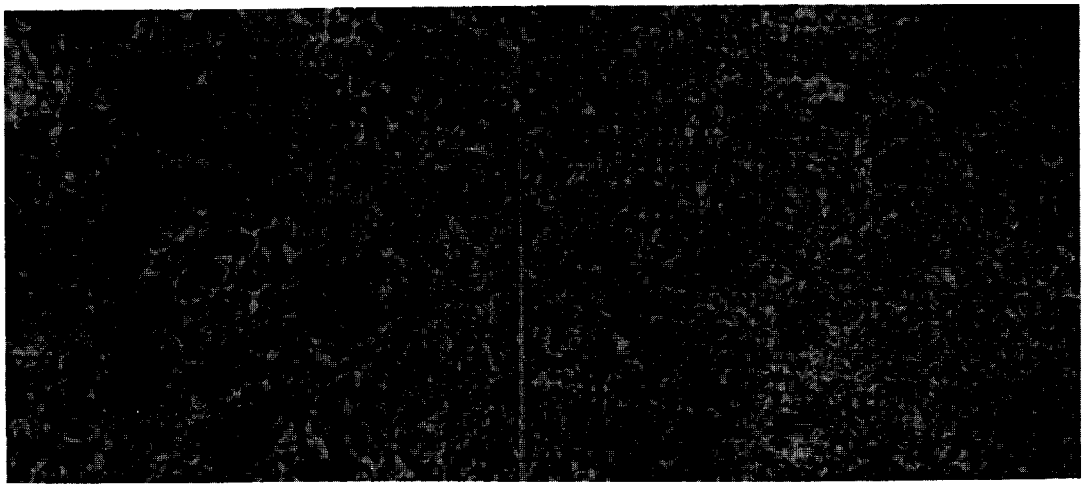
The photomicrographs of Fig. 32 (a and b) show the structure of this weld junction at 200X, where the properties are analogous to those of the junction between ductile iron and steel.

The interpenetration of the two metals along the length of the contact surfaces was extremely thorough, especially in the ferritic structure zones of the malleable iron, which blended uniformly into the ferritic structure of the steel.

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(a) Unetched.



(b) Nital etch.

malleable iron weld zone ductile iron

Fig. 30. Structure of the weld between malleable iron and ductile iron (200X).

Welds Between Nonferrous Base Metals With Ferrous Metals

The considerable difference between the physical and mechanical properties of nonferrous metals, such as copper (Cu), brass, or aluminum (Al), and those of ferrous metals such as the cast irons and steel, were studied with the objective of recording the welding parameters and establishing the methods of carrying out the operation itself in the most satisfying manner possible.

An optimum choice of welding parameters, since they do not correspond to the distinctly different characteristics of the different metals, is extremely difficult, since it is necessary to satisfy at the same time different requirements.

The difference in melting points is considerable between Cu, and especially between brass and Al, and the ferrous alloys (cast iron and steel). In the course of flash welding, the consumption of a nonferrous metal is greater than that of a ferrous metal.

The thermal conductivity of a ferrous metal is about 1/7 that of Cu and about 1/3 that of brass, with an intermediate value for Al. The high thermal conductivity of nonferrous metals, especially Cu and Al, makes it difficult to concentrate the heat within a relatively narrow zone, with the consequent extension of the injurious effects produced by the thermal cycle in the part.

To increase the thermal gradient when metals characterized by a high thermal conductivity are present, it is possible to vary three parameters, namely the speed of flash welding, the depth of flashing (that is, the consumption of metal), and the final distance of the jaws. This final distance has an influence inversely proportional to the temperature gradient.

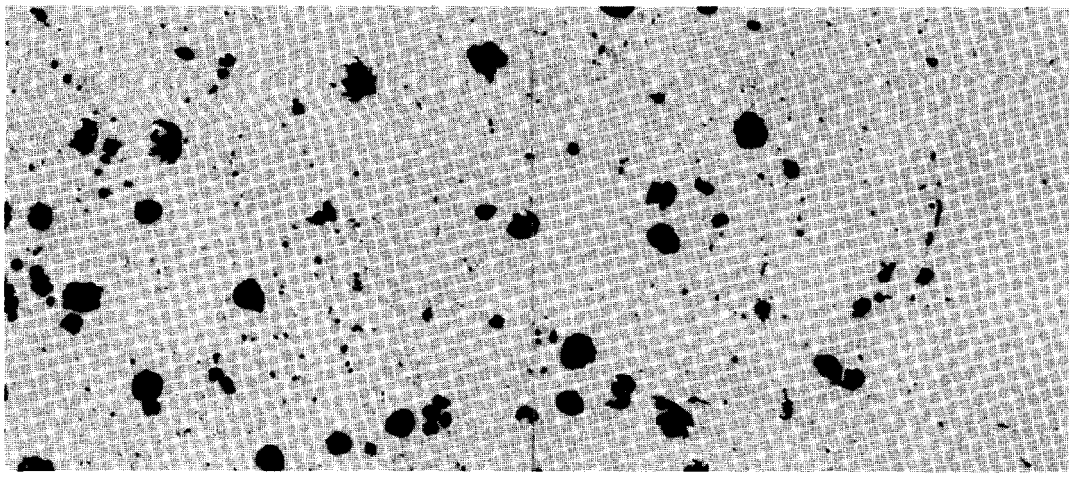
The thermal gradient increases with the speed of flash welding, since at high speeds the accumulation of heat is more significant.

With small metal consumption, all the other parameters being equal, a systematic (regular) situation cannot be attained since, due to the high thermal conductivity, an excessive loss of heat takes place. Because of this, in order to permit a sufficient concentration of heat so as to produce the weld union, the overhang of nonferrous metal with respect to the welding machine jaw should be greater than that of the ferrous metal.

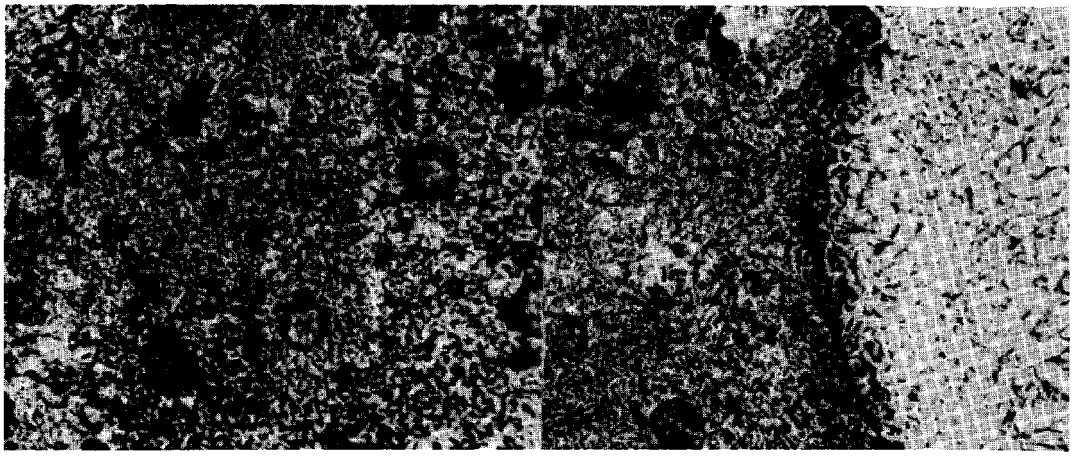
In Table 9 are reported some geometric variables with relation to the diameter of test specimens used during the welding of various metals.

Over and above an appropriate pressure on the jaws, it is equally necessary to change all the electrical and mechanical parameters of the welding cycle so as to attain a sufficient concentration of heat without causing an excessive consump-

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(a) Unetched.



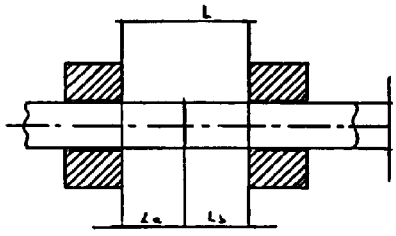
(b) Nital etch.

ductile iron weld zone steel

Fig. 31. Structure of the weld between ductile iron and steel (200X).

Table 9. Dimensional Changes in Welded Parts

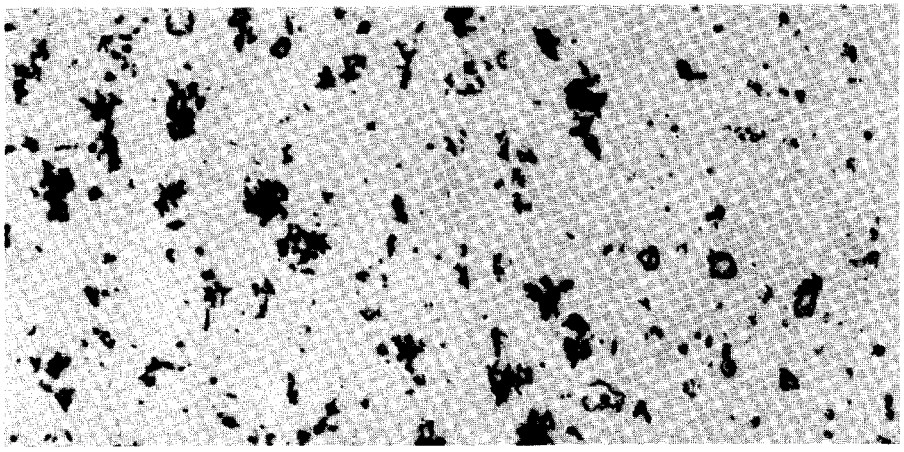
High carbon steel		La = 0.6d + Lb
Low carbon steel		La = 1.9d + Lb
Brass		La = 1.4d + Lb
Copper		La = 0.75d + Lb
High carbon steel welded with:		
Steel with low carbon %		La = 0.6d + Lb = 1.5d
Steel with low carbon % welded with:	*	La = 0.8d + Lb = 3.5d
Copper	*	La = d + Lb = 3d



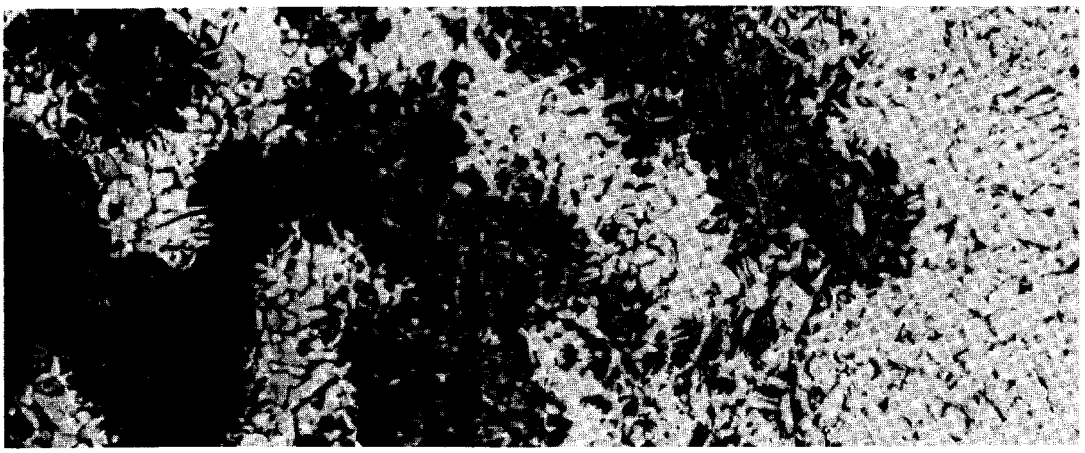
La = distance of each individual piece before welding
Lb = distance of each individual piece before welding *
d = diameter of pieces to be welded

* Trans. Note: These three entries seem to have errors. The third repeats the line above it, but should probably read "after welding".

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(a) Unetched.



(b) Nital etch.

Fig. 32. Structure of the weld between malleable iron and steel (200X).

tion of metal.

The hot mechanical properties, which vary greatly among the different metals, require control of the squeezing force as a function of the limiting mechanical strength of the softer nonferrous metal. This brings about a light forging of the ferrous metal and, as a consequence, the interpenetration between the nonferrous metal and the iron, rather than steel, can be foreseen, because of the difference in plasticity between the two types of ferrous metals at the working temperature. Some results have been obtained in welding ferrous metals to nonferrous metals by using very high squeeze pressures, with preparation of the test bar as shown in Fig. 33, with welding time and feed stresses appropriate to the nature and properties of the nonferrous metals as shown in Table 10.

The flash butt welding tests between ferrous and nonferrous metals were carried out between the following metal types:

Cu type Cu DHP welded with:
- Flake graphite gray cast iron type G 20 UNI 15007/69
- Ductile iron type GS 55/5 UNI 4554
- Malleable cast iron type GMN 35 UNI 3779/69
- Steel type Fe45 UNI 3596

Brass type 60 40 UNI 4892 welded with:
- Flake graphite gray iron
- Ductile iron
- Malleable cast iron
- Steel

Al UNI 4507 welded with:
- Flake graphite gray cast iron
- Ductile iron
- Malleable cast iron
- Steel

Cu Type Cu DHP

The work was carried out with a commercial Cu, type Cu DHP, with a Cu content in excess of 99.80%, deoxidized with phosphorus, with a phosphorus content of 0.013 to 0.050%, rolled into rods 30 mm in diameter and having the following mechanical properties:

Rupture strength = 20-24 kg/mm^2
Yield strength, 0.2% = 4 - 6 kg/mm^2
Elongation, % on 5 diameters = 45 - 50
Brinell hardness HB 10/1000/30 = 45 - 50.

The coefficients of thermal expansion are slightly different for the Cu-steel weld, while the difference is significant for the Cu-cast iron weld; so that in cooling, internal stresses might arise which, if they are not immediately controlled, could cause rupture of the weld and reduce the properties of the junction.

The solubility of Cu in iron at temperatures above 910C (1670F), that is, in the austenitic phase, can reach 9% Cu in the iron, while below these temperature values it is extremely limited and can give rise to globular Cu inclusions in the ferrous metal. The solubility of Cu in iron is practically nil in the solid

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state, for which reason at temperatures below 900C (1652F) the two metals should be considered entirely immiscible. At this temperature they form two clearly distinct phases without the formation of zones of intermediate composition or solid solutions, so that the soundness of the weld will be determined predominantly by the manner in which efficient squeezing takes place and by extensive and uniform interpenetration.

There are several ways to increase the thermal gradient: by increasing the amperage, by using a higher welding speed, or by preparing the parts as shown in Fig. 33. Using a higher welding speed will increase the consumption of metal during the flash cycle. Table 11 reports the mechanical properties obtained in tests carried out on the test bars welded between Cu and the metals indicated above. The test bars all broke along the length of the weld surface at a rupture strength always weaker than that of Cu, from about 10 to 17 m-kg/mm^2. Nevertheless, this is greater than the permanent deformation limit of Cu, which permits elongations of 2 to 10% over half of the test bar on the Cu side.

Lower elongations were also observed in the halves of the test bars on the ferrous metal side, ranging from 0.6 to 1.3% for the ductile iron and malleable iron and up to 4% for steel, which demonstrates some degree of toughness in the weld. A strong weld is also confirmed by the high impact strength values and is particularly noticeable in the welds between Cu and malleable iron and Cu-steel, which leads us to think that the weld is stronger if it is made between Cu and ferrous metals with a predominantly ferritic structure, instead of with ferrous metals having a high percentage of combined carbon.

The fracture cross section shows a gray background with round spots having the appearance of flakes (nodules) of copper, which denotes poor uniformity.

The hardness diagrams taken longitudinally to the joint are shown in Fig. 34. An increase of hardness is found to correspond to the weld zones. On the one hand, hardness drops off rapidly a short distance from the weld line. There is a progressive decrease in hardness until the minimum value for each ferrous base metal is reached. The hardness values, in the case of cast iron, at 1 mm from the weld, are comparable with the same measurements made on test bars of these metals welded between themselves.

The soft steel does not show any increase in hardness in the zone close to the weld, which confirms that there is no formation of hard structures.

Table 10. Voltage, Pressure and Time Values Used in Welding Nonferrous to Ferrous Metals

	Feed Voltage	Pressure kg/cm^2	Weld Time Sec.
Cu + GG	380 V	P = 8	t = 10 — 15
Cu + GS	380 V	P = 8	t = 10 — 15
Cu + GMN	380 V	P = 8	t = 10 — 15
Brass + GS	380 V	P = 7	t = 10 — 15
Brass + GMN	320 V	P = 7	t = 10 — 15
Brass + Fe 45	320 V	P = 8,5	t = 30
Cu + Fe 45	320 V	P = 10	t = 30

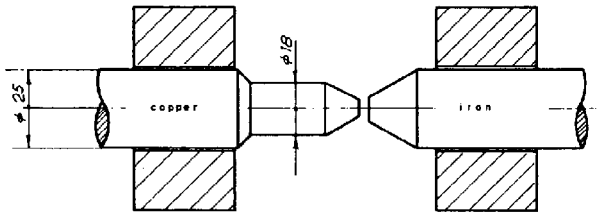


Fig. 33. Arrangement for welding nonferrous metals to iron using very high squeeze pressures.

The structures formed in the contact zone between copper and the different types of cast iron are very similar. Figure 35 shows the weld zone between Cu and flake graphite cast iron. Nital etch at 200X. In the contact zone between Cu and the different types of cast iron, a decarburized layer is formed. A considerable diffusion of Cu droplets was noted in the body matrix of cast iron which, in the black and white photomicrograph, appears as round or oval globules with a uniform gray color, while in reality they have a bright copper-red color.

Table 11. Mechanical Properties of Test Bars of DHP Copper Welded to Various Ferrous Metals

		Gray Iron		Ductile Iron		Malleable Iron		Steel	
Dimensions	d	15	—	15	—	15	—	15	—
	b	—	10	—	10	—	10	—	10
	a	—	8	176,7	8	—	8	—	8
Section mm^2		176,7	80	75	80	176,7	80	176	80
Useful part, mm		75	40	1700	40	75	40	75	40
Total strength, kg		1800	—	9,6	—	2550	—	3100	—
Unit strength, kg/mm^2		10,4	—		—	14,4	—	17,5	—
Elongation, % on 5 D		1,2	—	1,8 Cu 0,6 GS	—	4,2 Cu 1,3 GMN	—	10,1 Cu 4 steel	—
Impact strength, kg/cm^2		—	17,5 44,5	—	31,9 18,7	—	81 137	—	156

Observations: All specimens broke in the weld/

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Table 12. Tensile Strength and Impact Strength of Test Bars of Brass Welded to Various Ferrous Metals

		Gray Iron		Ductile Iron		Malleable Iron		Steel	
Dimensions	d	6	—	9	—	6	—	15	—
	a	—	10	—	10	—	10	—	10
	b	—	8	—	8	—	8	—	8
Section mm²		28,3	80	68,8	80	28,3	80	176,5	80
Useful part, mm		--	40	—	40	—	40	176,5	40
Total strength, kg		500	—	1850	—	1100	—	5650	—
Unit strength, kg/mm²		17,6	—	28	—	38,9	—	32	—
Elongation, % on 5 D		—	—	—	—	—	—	1% brass 0,8% steel	
Impact strength, kg/cm²		28,5 30	—	25 30	—	—	20 28,7	—	35 37,4

Observations: Rupture took place directly on the weld line in all cases.

On the Cu side, immediately after the contact surface, no transformation of any kind was observed. On the cast iron side, in the zone close to the contact surface, there was no graphite segregation and the formation of diversified structures or considerably altered structures took place over a long time, followed by ledeburite structures and by extensive cementite formations, probably due to a high rate of cooling. This leads us to believe that after the welding phase, the close contact of the cast iron with Cu, which is an excellent heat conductor, has accelerated the cooling of the welded part, thus simplifying the formation of unstable hard structures.

The photomicrograph of Fig. 36 shows the structure of the welded metals, Cu and steel, longitudinally to the weld zone, revealed by Nital etch at 200X.

There is no diffusion between the two metals joined and the line of separation between them in quite clear, also because of the absence of free carbon. There are no transformations in the steel structure, which remains entirely ferritic; this is equally confirmed by the development of the hardness, which remains constant from the weld surface for the entire length of the steel side. It shows only a modest enlargement in the dimensions of the ferrite grains.

Brass Type 60-40 UNI 4892

The mechanical properties of brass 60-40 UNI 4892 are as follows:

Rupture strength = 36-38 kg mm^2
Yield strength, 0.2% = 14 kg mm^2
Elongation, % on 5 diameters = 45-50
Brinell hardness, HB 10/1000/30 = 75 - 80

This type of brass is characterized by a two-phase structure, $\alpha + \beta$ and it is available commercially in the form of semifinished products such as tubes, bars, sheets, and wire.

The welding of brass to ferrous metals causes problems similar to those of Cu, although in a smaller measure, so that because of its considerably lower thermal conductivity (around half that of Cu), the brass presents a greater opportunity to concentrate the heat in the weld zone. Counteracting this positive aspect, it has a melting point lower than that of Cu. The large difference between the melting points of brass and those of the ferrous metals should be compensated for by a greater consumption of brass during the flash welding.

The mechanical properties of brass at elevated temperatures are not as good as those of steel and cast iron, so that the squeezing pressures should be regulated in such a way as to avoid cracks and significant deformations in the weaker metal. As we have seen, Cu has practically no solubility in iron. On the contrary, zinc in the brass alloy has a significant solubility in α iron at high temperature which, however, diminishes rapidly with temperature. As can be deduced from an examination of the Fe-Zn phase diagram, at low temperatures a homogeneous solution of α iron and zinc is in equilibrium with an intermediate phase of intermetallic composition, $FeZn_7$. This compound, which is hard and brittle, gives the Fe-Zn alloy poor mechanical properties.

We must also take into account the fact that zinc, which is found in brass alloys in a percentage lower than that of Cu, is subject to a certain loss during the flash welding cycle, because of its great volatility at elevated temperatures. As a consequence, its diffusion into the iron will be limited by this and also because of the low concentration gradient. Table 12 summarizes the results obtained in the tensile and impact strength tests on flash butt welded bars between brass and ferrous metals.

Unfortunately, the material available did not permit us to obtain for some tests any other than test specimens of very reduced dimensions, with which it was nevertheless possible to determine the tensile strength.

Rupture of the welded test bars always took place the length of the welded surface, with a fracture of yellow color, slightly and uniformly granular.

The rupture strength in the different tests is notably superior to that of tensile strength tests on test bars welded between the corresponding ferrous metals and copper. It reached the strength value of brass in the test bars welded between brass, malleable iron and steel, while the test bar welded with gray iron broke at a load close to that of the latter metal.

Elongation of a low value took place only in the welded test bar between brass and steel.

The values of the impact test were good.

Figure 37 shows schematically the hardness values longitudinal to the weld line from one to the other base metal. In contrast with the case previously examined, the maximum hardness values were not found opposite the weld line, in the ferrous metal, but at some millimeters distant from the weld. This zone is characterized by the presence of a martensitic

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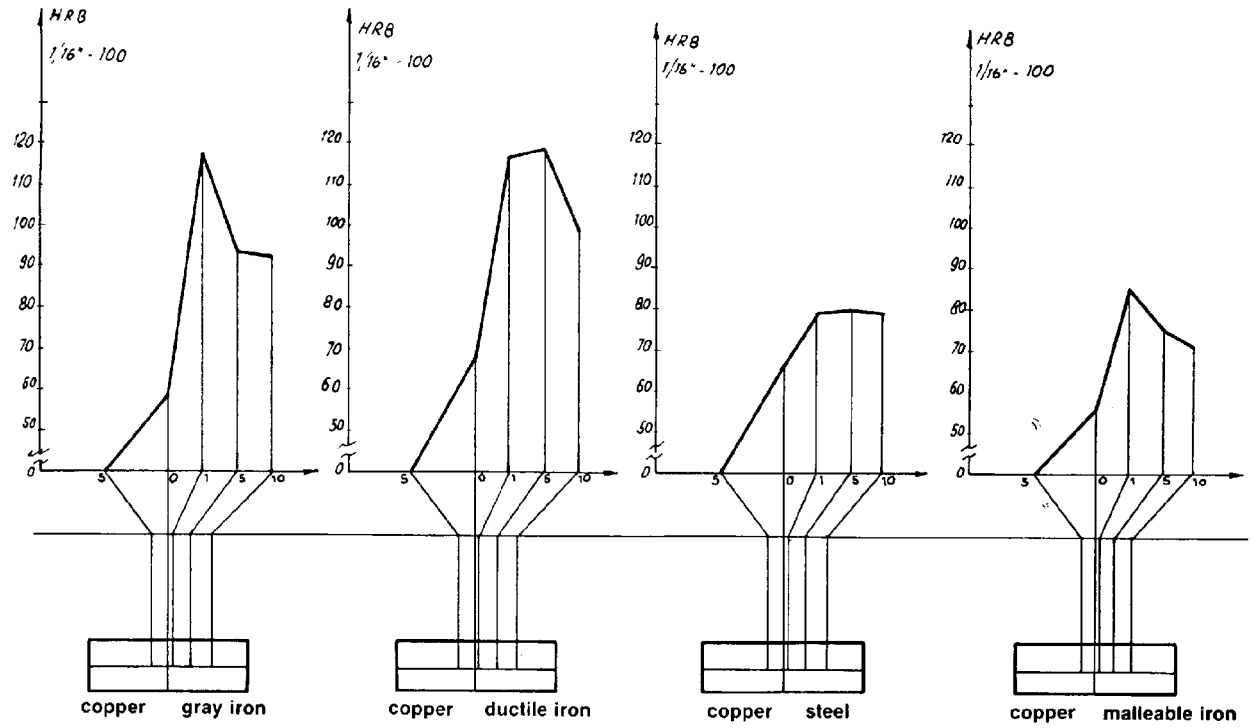


Fig. 34. Hardness diagrams of welds between copper and various ferrous metals.

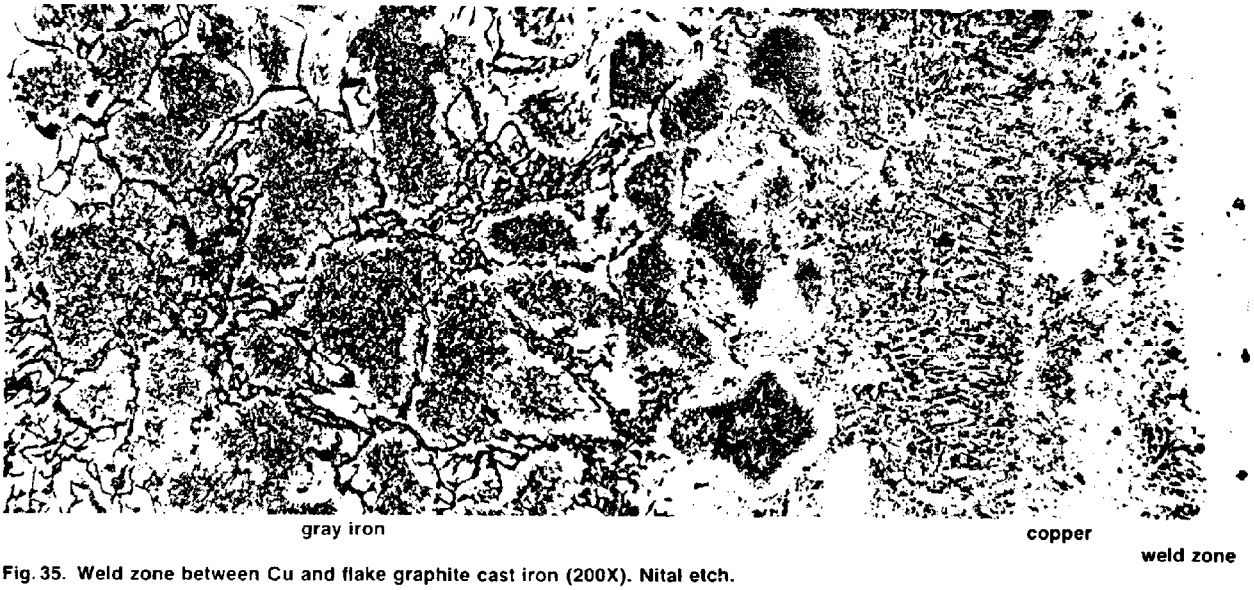


Fig. 35. Weld zone between Cu and flake graphite cast iron (200X). Nital etch.

structure with cementite, which is formed in the course of rapid cooling after welding. Going beyond this zone, the hardness values diminish rapidly until they reach the values typical of the unaltered base metals at a distance of 5 to 10 mm from the junction line.

The structural appearance of welds between cast irons of various types and brass shows similarities to all the cases studied. Figure 38 shows the structure of the weld zone between brass and ductile iron, revealed by Nital etch for the cast iron and NH_3-H_2O_2 etch for the brass, at a magnification of 750X.

On the whole, considerable penetration of the brass into the ferrous metals was noted in the form of large and small drops and large blades of brass extended deeply into the matrix of the cast iron. In order to explain the considerable dispersion of the brass, we can think of intergranular diffusion phenomena similar to those which take place in brazing techniques for ferrous metals with brass. Actually, it is well known that for the brazing of steel parts and cast iron, Cu is used as well as brass. We can then visualize a connection by interpenetration of liquid brass into the mass of ferrous metal.

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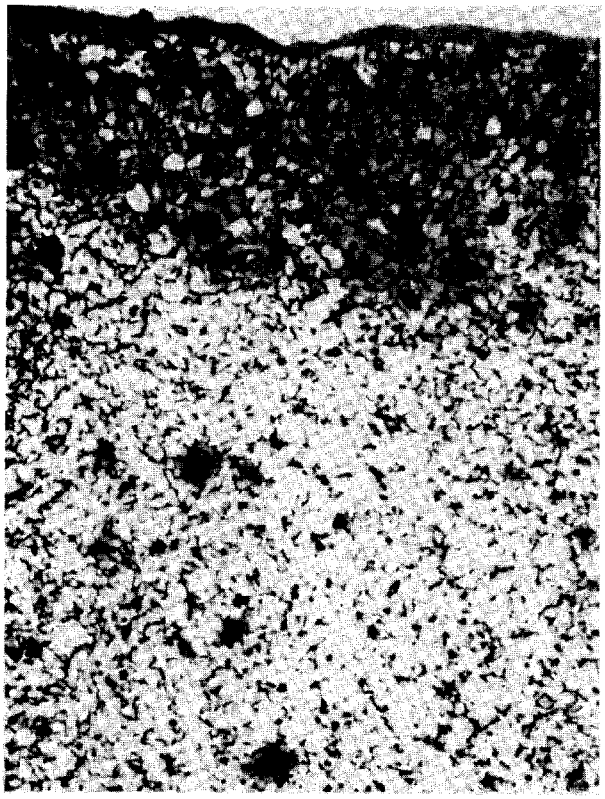


Fig. 36. Longitudinal section showing the structure of the weld between Cu and steel (200X). Nital etch.

The structures in the thermally affected zone on the ferrous metals side reveal the presence of definite martensitic formations in the layers close to the contact surfaces. Beyond a certain depth, these martensitic formations are mixed with carbides and with brass penetrations, confirming the development of the hard areas shown in Fig. 37.

Figure 39 shows the structure of the weld zone between brass and steel as revealed by Nital etch for the steel portion and NH_4-H_2O_2 for the brass portion, at 200X. In the brass there is a mixed structure of α-β phases with the latter predominating as a consequence of the thermal cycle of welding. The steel has not undergone any structurally significant modifications.

Aluminum UNI 4507 with Ferrous Metals

Flash welding of Al is extremely difficult due to the specific physical and mechanical properties peculiar to this metal. Its very low melting point of 660C (1220F) requires a very close control of the electrical parameters while its mechanical properties, which are very modest at welding temperature, impose a rigorous limit on the mechanical parameters. Its high thermal conductivity makes it difficult to concentrate heat in a relatively narrow zone, which involves an extension of the undesirable effects which welding produces in the part. Flash butt welding between Al and ferrous metals involves insurmountable difficulties because of the large difference between the physical and mechanical properties of these metals. Actually, in the tests which we carried out, we did not succeed in producing true welds successfully, but only "pasting" or "glueing" the specimens together, which did not have any strength.

Only for very limited sections were we successful in preserving some welded test bars, from which it was possible to make only impact strength test bars and specimens for microscopic

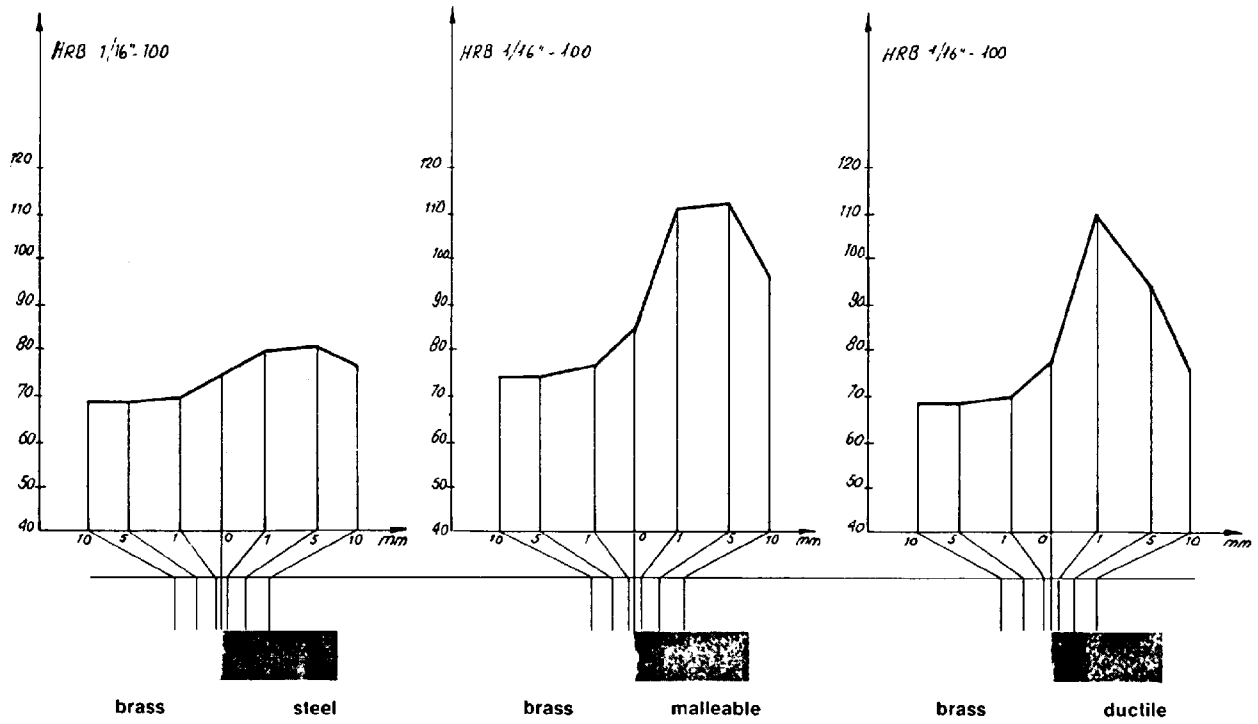


Fig. 37. Hardness values of welds between 60-40 UNI 4892 brass and various ferrous metals.

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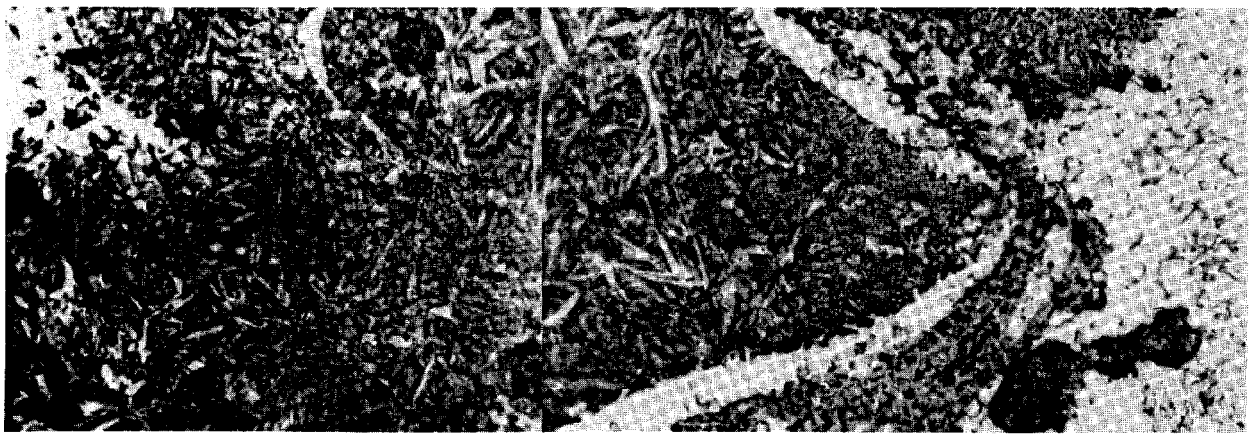


Fig. 38. Structure of weld zone between brass and ductile iron (200X). Nital etch.

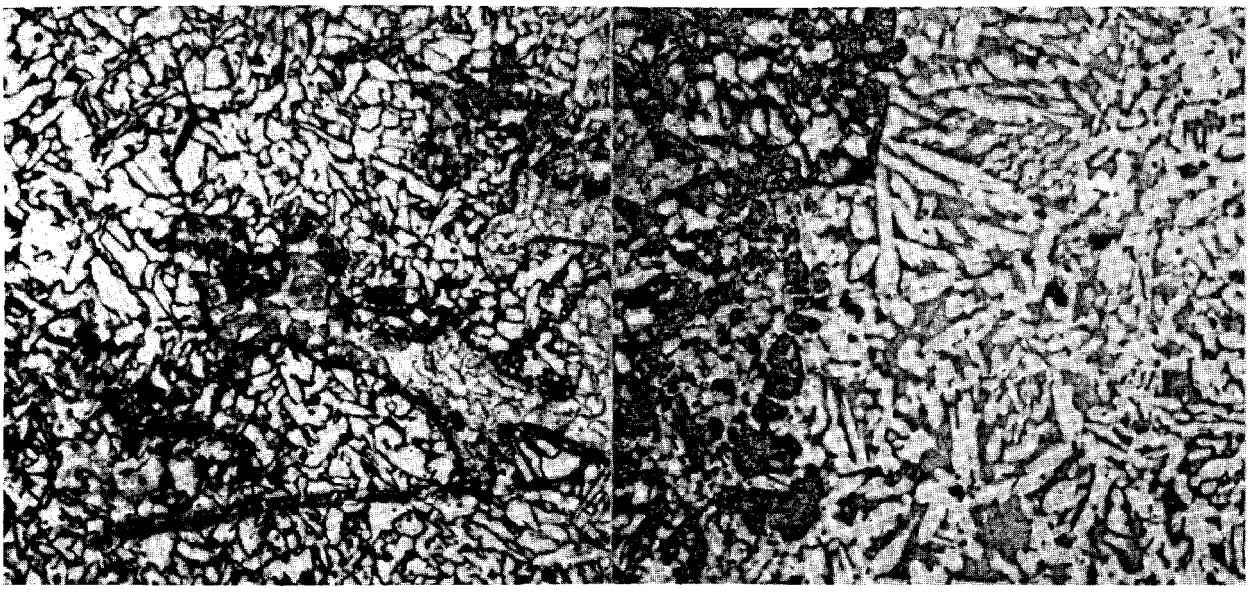


Fig. 39. Structure of weld zone between brass and steel (200X). Nital etch on steel, NH_3-H_2O_2 etch on brass.

examination, since the welded test specimens of any significant section size always broke in the welded section on simply falling to the floor or during machining for preparation of the test specimen.

Figure 40 shows the hardness diagram longitudinal to the welds which, in a way similar to all the cases studied previously, presents maximum hardness points in the thermally influenced zones of the ferrous metals near the surface of the weld. These diminish down to the typical value for each ferrous metal, while on the Al side the hardness drops off immediately to the values characteristic of this metal.

The inconsistency of these welds is shown equally by the minimum values of impact strength which it was possible to carry out on some test bars which we succeeded in machining.

Al + flake graphite cast iron	0.75 kg/cm^2
Al + malleable iron	0.75 kg/cm^2
Al + ductile iron	0.70 kg/cm^2
Al + steel	0.65 kg/cm^2

Figure 41 shows the structure in the weld zone between Al and ductile iron with Nital etch at 750X. Wide, clear zones are revealed which lead away from the Al and penetrate the ductile iron, but in spite of the deep interpenetration of the two metals which were joined, the strength of the weld is entirely inconsistent.

Around the globules of graphite there is the usual partial solution of free carbon and its diffusion into the surrounding zones, with formation of a martensitic structure in the zone next

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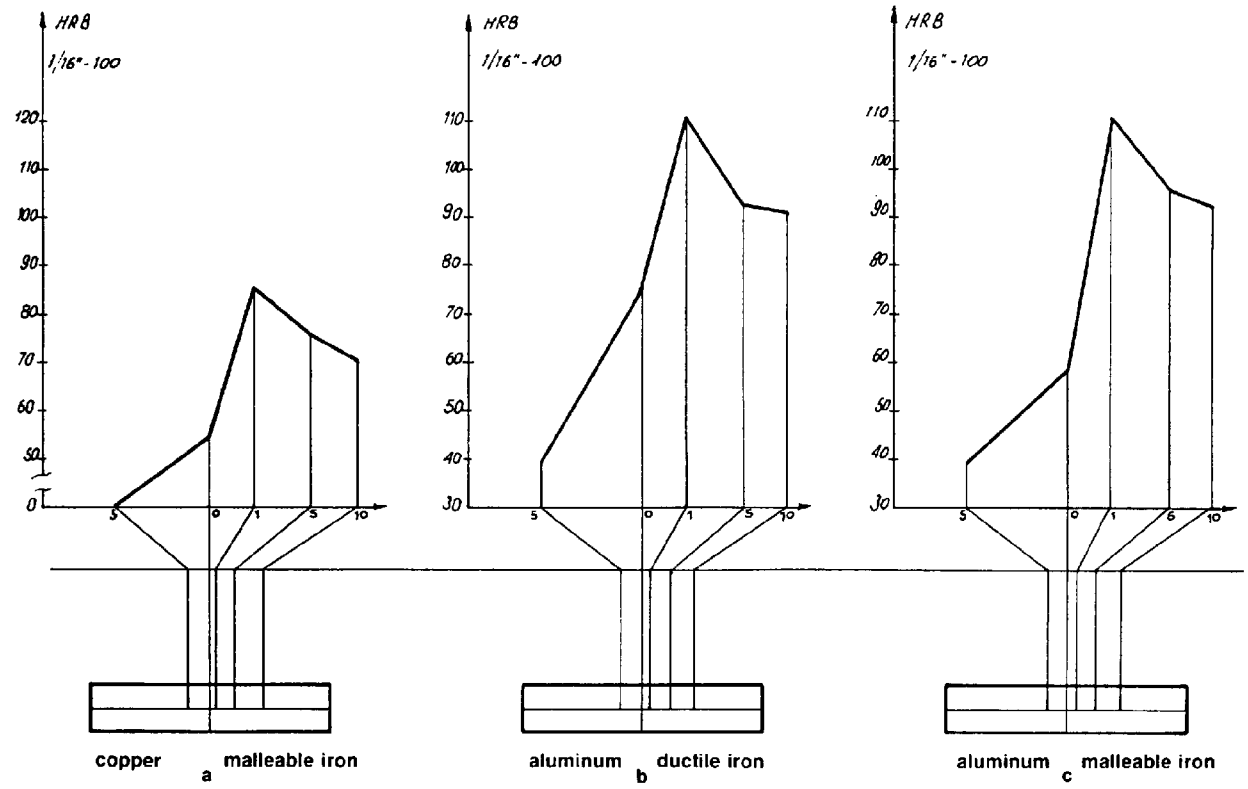


Fig. 40. Hardness values of welds between Cu, Al and various irons.



Fig. 41. Structure of weld zone between Al and ductile iron (750X). Nital etch.

to the contact surfaces between the two metals. This explains the peak of hardness observed.

The contact zone often presents welding defects, the presence of oxide inclusions and continuous zones of metal solid solutions, which would tend to be weak and brittle. This explains the lack of strength of the weld.

Conclusions

From the experimental results described and explained the following conclusions can be made.

Flash Butt Welding

Ferrous Metal Base (Between Two Homogeneous Cast Irons)

1) All types of cast iron can be easily and totally welded between themselves by the system described. The welded surface tends to be completely uniform, with good continuous interpenetration. There are no continuous zones of metallic solid solutions, which would tend to be weak and brittle.

2) The mechanical properties of the welded joint, the rupture and impact strength, are all comparable with those of the base

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metal and, as a consequence, all the types of cast iron appear to form welds as strong as the original base metal.

3) The structures show a progressive transformation toward both sides of the weld contact surfaces, starting from the unaltered base metal, with an enrichment of finer and finer pearlite as we proceed toward the weld. A gradual increase, however limited, of the hardness occurs as the weld is approached. The maximum hardness value is obtained exactly at the weld itself.

4) With the exception of austenitic ductile iron, scattered hard structures of ledeburite or cementite form in the weld contact zone, which corresponds to the greatest hardness measured.

5) A short anneal after the welding cycle, carried out on the welding machine itself, eliminates all of the ledeburitic structures and even the network of cementite which is formed in the junction zone, transforming it into an entirely pearlitic or troostitic structure which is strong and tough.

6) The graphitic segregations are expelled from the surfaces to be joined in the course of flash welding. They are partially dissolved in the zones next to the weld section, especially when dealing with graphite of the flake and floccular type, leaving in the junction zone no more than a few globules of graphite which contribute to an improvement in strength. Only the graphite of spheroidal type remains practically unaltered, which illustrates a difference in behavior between the different types of graphite with respect to the thermal influence of the welding cycle.

This leads us to believe that there exists a difference in the tendency for graphitic carbon to go into solution and to diffuse in iron according to its morphological shape and the extent of its specific surface which is in contact with the ferritic matrix. This seems to be more significant for graphite of the flake type, of intermediate importance for graphite of the floccular (malleable nodule) type and of minimum importance for the spheroidal type graphite.

7) The thermal cycle of welding, characterized by rapid heating up to a temperature close to that of the melting point of the metal and followed by rapid cooling (Fig. 3), can be considered as a heat treatment or as a dissolving and diffusion of graphitic carbon into the iron. This enriches the iron in combined carbon and when followed by a quench treatment, facilitates the formation of hard structures (ledeburite, cementite and, eventually, martensite). If to this quench treatment we follow up with an anneal treatment for a short but sufficient time, we can dissolve the hard elements which have formed following the quench and produce an equilibrium structure, stress-relieved, which is tough and strong. This equilibrium structure will be of troostite or of pearlite.

Between Heterogeneous Ferrous Metals

1) The different types of cast iron are easily and completely weldable, one to the other, even with steel.

2) The difference in welding parameters for the different combinations is not significant, taking into account the good uniformity of mechanical and thermal properties. The only difference for welding with steel consists of getting suitable grip for the jaws.

3) The surfaces of the welds turn out to be perfectly uniform and continuous. They are characterized by a strength level generally greater than that of the weaker metal. That is why the junctions resist stress without cracking and the fractures of the tensile test bars take place in the weaker of the two metals joined, at a rupture strength which is characteristic of that metal.

The impact strength values are generally good, even if they are occasionally affected by the hard structures which may form in the junction zone.

4) In a similar way to that stated before, in the weld zone between two different types of cast iron structures of ledeburite or scattered network cementite often form, which correspond to a high hardness value. On both sides of the weld, the structures are progressively transformed because of phenomena already described, namely the progressive solution of the free carbon into the iron and the formation of structures with combined carbon in such a way that troostite or pearlite are formed. Mechanical properties even better than those of the base metal are thus found.

5) In a similar way to that stated above, a brief anneal carried out within the welding cycle itself eliminates all hard structures of ledeburite and network cementite formations by favoring the formation of an entirely pearlitic or troostitic structure. This structure is strong and tough, even in the welds of heterogeneous ferrous metals.

6) This experimentation also confirms what has been extensively explained above relative to the behavior of different types of graphite in the thermal cycle of welding.

Nonferrous Base Metals with Ferrous Metals

1) The welds between these metals, which have very different physical characteristics, are extremely difficult to make and require careful control and adaptation of all the welding parameters. The factors which most affect the welding are the different thermal conductivities and the difference in melting points.

Adequate preparation of the parts to be welded and exact positioning in the welding machine are necessary for a good weld.

2) The strength of the welds is always less than that of the two metals joined, but always close to that of the weaker metal. The fracture takes place in the junction zone. The values of impact strength are generally good.

3) In the weld zone and in the surrounding zone, on the ferrous metal side, all of the structural transformations take place as indicated previously.

In Cu there is no structural transformation except for a slight tempering effect.

In brass there is an increase in β phase with consequent hardening of the metal and an increase in its strength characteristics.

4) Flash welding between Al and the ferrous metals, such as cast iron and steel, is practically impossible because of the excessive difference in melting points of the metals.

Even with the best possible proportioning of the several electrical and mechanical parameters of welding, the weld lacks any mechanical strength.

Possible Applications

What industrial applications can be found for this system of joining cast iron parts, or cast iron with metals of a different nature, whether ferrous or nonferrous, taking into account the fact that it must be carried out on extremely costly special machinery with an economic yield only for work in very long mass production runs?

Consider that with different types of malleable iron parts are produced for use in the automotive industry, such as: crankshafts, motor mounts, differential carriers, brake drums, steering boxes, pedals and steering knuckles; in the motorcycle industry, such as: pipes for wheel forks and chassis and motor cylinder blocks; in the bicycle industry, such as: wheel hubs, chassis tubing and pedals; in the tractor industry, in the construction of agricultural machines and sewing machines.

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We also note that the experiments in the flash butt welding of iron have not taken place anywhere except in Russia and for very specific applications, such as the welding of cast iron pipe. Nothing is found in the technical literature of other countries.

Among the most likely applications, there is the possibility of welding hollow bodies in gray iron or ductile iron for use in flasks with shaped, curved ends, T-elements or others at the end of pipes which are presently centrifugally cast, and where the system of fabrication does not permit the production of shaped ends. Solid bodies of round, square or other sectional shapes might be welded by these techniques.

Nowhere has the problem of flash welding of different metals ever been approached at the research level, even though there probably exist industries which could use it profitably.

Actually, the operating conditions of a particular component part of a machine are not always uniform in all sections. If two zones of the same component part which must satisfy different demands are produced with the intent of using a particular metal which meets the different imposed conditions, it would doubtless be possible to effect significant metal economies by improving the overall properties of the component part.

Sometimes a part to be welded calls for a combination of properties that are metallurgically incompatible. An example of this might be a part that has to have high corrosion resistance and also good mechanical properties.

Thus, for example, welds made between pieces of different ferrous metals, such as cast iron and steel, including stainless steel, can find certain practical applications. Welding between parts of dissimilar metals such as Cu and brass with ferrous metals can accentuate the difference in their colors and find a practical application among decorative objects. Also, because of the different strength and corrosion resistance of certain chemical or natural agents, it could be used in the joining of piping for chemical or petrochemical equipment, in the construction of tubular channels which must resist certain corrosive actions and, as a consequence, should be fabricated partially in cast iron, steel, stainless steel, or special brasses, with their components in non-special metals positioned outside of the area of contact with corrosive elements.

It is easy to foresee that when engineers informed of the availability and the validity of this type of weld, they will find it to be the solution to a great number of their problems.

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Production Welding of Cast Irons

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ABSTRACT

The use of cast irons for welded components has been severely restricted by the lack of a welding wire that would produce high-quality welded joints in the materials by an automatic process. Merely because of welding requirements, forged or cast steel has been used for many components that otherwise could have been made of a cast iron. The use of economical cast irons for such parts could substantially reduce production costs for many types of assembled equipment. A recently developed flux-cored wire now makes it possible to extend the benefits of cast irons to welded structures.

Many attempts have been made to develop a welding wire for cast irons but until recently efforts were thwarted by the low ductility of heat-affected zones in welded cast irons. Shrinkage stresses during cooling of the weld metal often caused heat-affected-zone cracking. It has long been known however that cast irons could be successfully welded with nickel-alloy weld metal having a high (about 1%) carbon content. The low solubility of carbon in nickel results in precipitation of graphite during cooling of the weld metal. The resulting volume increase holds shrinkage stresses to a low level and prevents heat-affected-zone cracking. The graphite-rejection principle is the major reason for the success of nickel-alloy coated electrodes (AWS classes ENi-CI and ENiFe-CI) for manual welding of cast irons.

With coated electrodes, carbon can be added to the weld metal through the flux coating. A conventional bare wire for automatic welding cannot feasibly be manufactured with the required high carbon content. With flux-cored wire however the graphite-rejection principle can be applied to automatic welding through carbon added to the flux core. The flux-cored wire for cast irons produces high-quality welded joints having favorable heat-affected zones. The wire can be operated at travel speeds up to 100 in./min (254 cm/min) and provides deposition rates of up to about 18 lb/hr (9.2 kg/hr).

Introduction

Cast ductile, malleable and gray irons are economical, highly serviceable materials used for numerous components of assembled equipment such as highway vehicles, farm machinery, construction equipment and machine tools. Throughout the years, however, the use of cast irons has been impeded by the lack of welding products capable of creating high-quality welded joints in the materials. The problem has become especially acute with the development of modern manufacturing technology in which high-speed, automatic welding is integrated with the assembly sequence. Because of the welding requirements, designers are often forced to specify expensive forged or cast steel for parts that would deliver equal performance in cast iron. The substitution of cast iron for such parts would save literally millions of dollars in many industries. The opportunity to gain those savings is now presented by a new flux-cored welding wire.[1]

Several factors influence the achievement of satisfactory welds in cast irons. The deposited weld metal must have high strength and ductility. Because of the comparatively low ductility of most cast irons, the shrinkage stresses developed during cooling of the weld deposit can cause the base material to crack in the heat-affected zone (HAZ). The cracking problem is compounded by metallurgical changes in the HAZ that further decrease ductility in that area.

Most of the problems with manual welding of cast irons were solved over twenty years ago with the introduction of high-nickel (Ni)-coated electrodes especially formulated for cast irons.

The electrodes, AWS classes ENi-CI and ENiFe-CI, deposit either Ni or Ni-iron weld metal with a high (about 1.5%) carbon (C) content. The high-Ni weld metal provides the necessary strength and ductility. The high C content, which is added to the deposit through the flux coating, serves to decrease shrinkage stresses in the welded joint. Carbon has a low solid solubility in Ni and any excess C present in the molten weld metal will be rejected as graphite during solidification. The graphite increases the volume of weld metal and thereby decreases the amount of shrinkage and accompanying stress.

The Ni-alloy-coated electrodes that incorporate the graphite-rejection principle produce high-quality welds in cast irons. However, the low deposition rates and manual operation restrict use of the coated electrodes to repair welding and small-scale assembly operations. Large-scale production-line welding requires a filler wire that can be deposited at high rates by automatic processes.

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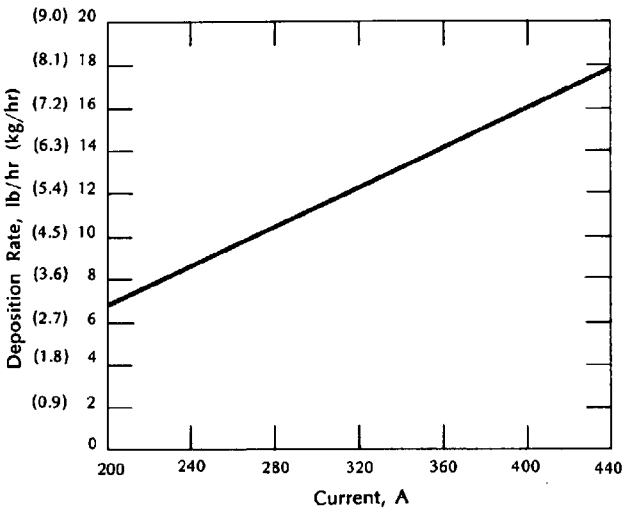


Fig. 1. Effect of welding current on deposition rate of flux-cored wire.

Although a need has long existed for a filler wire that would be equivalent to the coated electrodes, the very feature that accounts for the success of the electrodes prevents the manufacture of a conventional bare wire. A Ni or Ni-iron composition having the required high C content would be impossible to process by known methods.

The advent of flux-cored wire, however, presented the opportunity to combine the features of the coated electrodes with the high-speed welding of a bare wire. After much development work, a flux-cored wire having a Ni-iron sheath and a high-C core was successfully produced. Like the coated electrodes, the cored wire deposits sufficient C to promote graphite precipitation and lessen shrinkage stresses. In addition, the new wire produces a much more favorable HAZ which further decreases the possibility of base-metal cracking. The product is now in commercial use.[1] For the first time, high-quality cast-iron welds are being produced on the assembly line.

Welding Procedures

The new flux-cored wire is specifically designed for high-speed production welding of cast irons. The product consists of a Ni-iron tubular wire filled with C, slagging ingredients and deoxidizers. It is produced as standard 50-lb (23 kg) coils in diameters of 0.093 and 0.078 in. (2.4 and 2.0 mm). The nominal chemical composition of weld metal deposited by this flux-cored wire is 50% Ni, 1% C, 0.6% silicon (Si), 4.2% manganese (Mn) and 44% iron (Fe).

Welding with the cored wire is performed with standard equipment for flux-cored arc welding, usually the same equipment used for gas-metal-arc welding or submerged-arc welding. Direct-current power sources with constant potential and reverse polarity generally give the best results. A pulsing-arc power source has also been used successfully.

The core of the flux-cored wire contains deoxidizers and is designed to be operated without shielding gas. If welding conditions require greater protection, a shielding gas can be used with no effect on mechanical properties of the joint. Carbon dioxide is the most commonly used gas. The cored wire can also be operated with submerged-arc flux.

A wide range of welding travel speeds can be used with the flux-cored wire. High-quality welds have been produced at travel speeds from 10-100 in./min (25-254 cm/min).

The high deposition rates typical of automatic welding processes are readily achieved with this flux-cored wire. As shown in Fig. 1, deposition rates vary with welding current. Rates of 13-18 lb/hr (5.9-9.2 kg/hr) are obtained at normal welding currents. Such deposition rates are up to nine times greater than those obtainable with coated electrodes.

Preheat is not required for joints in ductile and malleable iron. For heavy-section, fully restrained joints in gray iron, preheating may be advantageous. None of the cast irons requires a postweld heat treatment after being welded with this wire.

Mechanical Properties

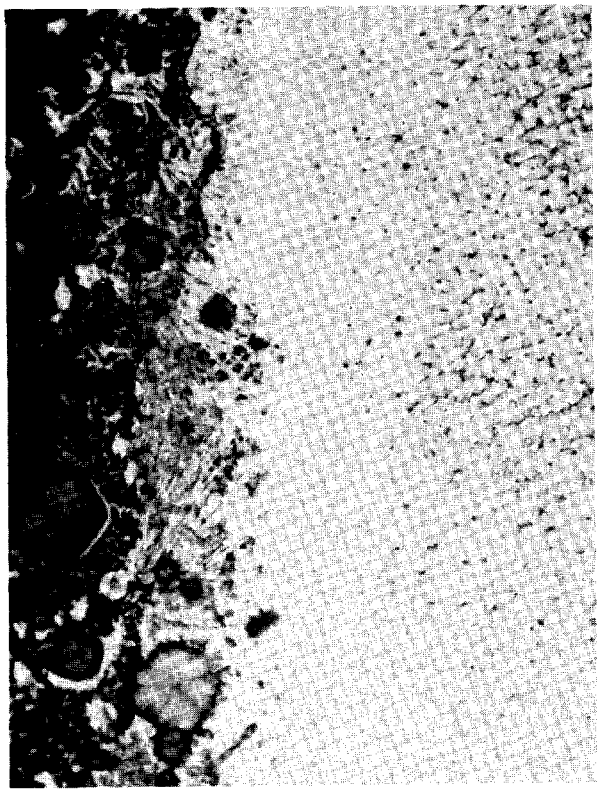
Joints in cast irons welded with this Plux-cored wire exhibit high mechanical strength. Table 1 lists typical mechanical properties of all-weld-metal samples and transverse samples from joints in

Table 1. Mechanical Properties of Joints Welded with the Flux-Cored Wire

Specimen	Shielding	Yield Strength (0.2% Offset)		Tensile Strength		Elongation,	Reduction of Area,	Hardness,
		ksi	MPa	ksi	MPa	%	%	Rb
All-Weld Metal	None	45.0	310	69.0	476	15.5	14.5	81
All-Weld Metal	CO_2	45.5	314	72.0	496	21.0	18.8	80
All-Weld Metal	Sub. Arc Flux	49.0	338	74.0	510	18.5	20.6	86
Transverse	None	43.5	300	66.0	455	-	-	-
Transverse	CO_2	44.0	303	66.0	455	-	-	-
Transverse	Sub. Arc Flux	45.0	310	64.0	441	-	-	-
All-Weld Metal*	CO_2	44.0	303	68.0	468	15.0	16.2	80
Transverse*	CO_2	43.5	300	67.7	467	-	-	-

*Pulsing-arc power source.

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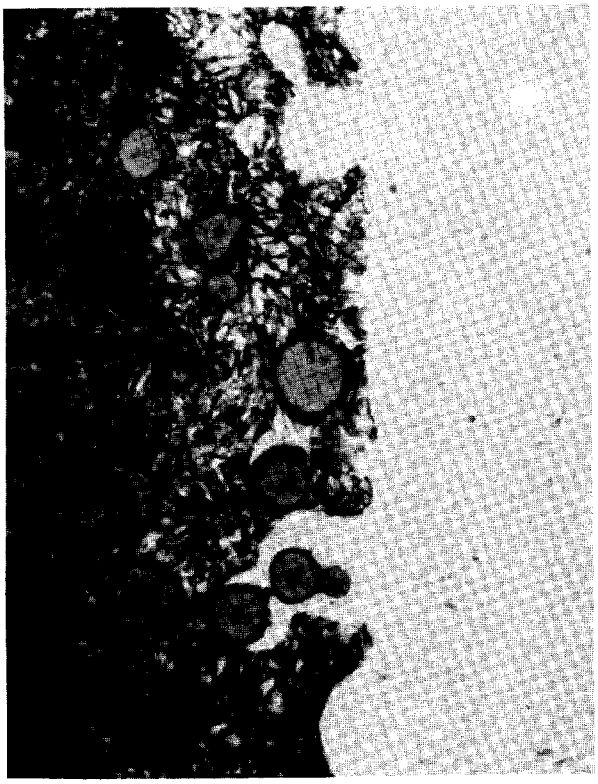


Fig. 2. Heat-affected zones from joints in ductile iron welded with, top) ENiFe-CI coated electrode and bottom) the flux-cored wire.

ductile iron. As indicated by the data, the tensile strength of weld metal deposited by the wire typically ranges from 68,000 to 74,000 psi (469-510 MPa), well above the 65,000 psi (448 MPa) minimum tensile strength of the most frequently used ductile iron (SAE grade D4512). In addition to their high strength, the weld deposits exhibit good ductility.

Heat-Affected Zone

The heat-affected zones of joints in cast irons have been the subject of much research and discussion. At one time, any welded joint in cast iron was suspect because of the formation of brittle carbides in the HAZ. The brittle carbides, coupled with the heavy shrinkage stresses of low-C welding products, often led to cracking problems. Now, however, metallurgists and design engineers accept the fact that all metals undergo metallurgical changes in their HAZs during fusion welding. The current approach for all materials, including cast iron, is to experimentally evaluate the effects of welding on service performance. For example, welded joints are often subjected to accelerated simulations of service conditions.

One of the advantages of flux-cored arc welding with the cored wire is an improved HAZ structure. Figure 2 compares the HAZs in joints welded with ENiFe-CI coated electrode and the fux-cored wire. The microstructure of the joint welded with the coated electrode contains a dense, continuous network of carbides along the HAZ. In contrast, the microstructure of the joint welded with the cored wire contains only sparse, discontinuous carbides in the HAZ, eliminating the possibility of a continuous band of brittle structure. The condition is further enhanced by the very irregular fusion line, which results from the high current density of flux-cored arc welding.

Mechanical tests confirm the absence of continuous brittle structures in HAZs of welds made with the cored wire. Figure 3 shows failure locations in transverse fatigue-test specimens. In none of the specimens does the path of failure follow the HAZ. Tensile tests produce similar results.

The effects of welding on service life should be evaluated for any welded structure. For most applications, however, the metallurgical changes in cast-iron joints welded with the cored wire are no more detrimental than those in welded joints in

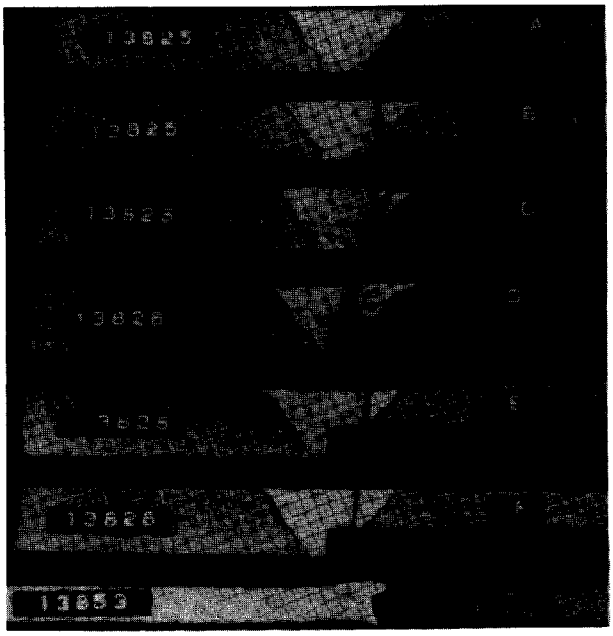


Fig. 3. Fatigue-test specimens from joints in ductile iron welded with the flux-cored wire.

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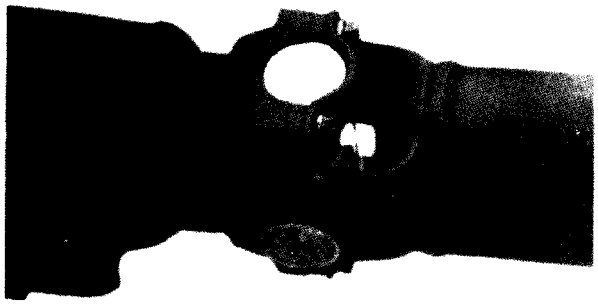


Fig. 4. Conventional universal joint composed of a cast iron yoke on the differential unit and a forged steel yoke welded to the drive shaft.

many low-alloy steels that are routinely used for welded construction.

Applications

Several manufacturers have begun programs to take advantage of the cost reductions available with the use of welded cast-iron components. The examples that follow are typical of the applications in which the cored wire is being used or evaluated.

Universal Joints

A typical universal joint in an automobile drive-shaft/differential assembly is shown in Fig. 4. The universal joint is composed of two different materials. The drive-shaft yoke, which must be welded to the carbon-steel drive shaft, is forged steel. The differential yoke, however, requires no welding and is a cast iron, in this case malleable iron. Cast iron obviously meets the requirements of power transmission since it is presently used for half of the universal joint. Steel is used for the welded half of the unit only because of the need to produce high-quality welds at high production speeds.

An extensive study conducted by one manufacturer has demonstrated the suitability of universal joints made entirely of ductile iron and welded with the cored wire. Figure 5 shows the welded joint in a prototype drive-shaft yoke of ductile iron. Prototype units have passed all life-cycle tests, including requirements for torsional fatigue strength and ultimate torsional strength. The redesigned unit can be produced at lower cost even though the ductile iron yoke weighs only 3-4 lb (1.4-1.8 kg).

Wheel Lock Bar

Figure 6 illustrates a lock bar used to lock the railroad wheels in position on a dual-purpose highway/railroad vehicle. The lock

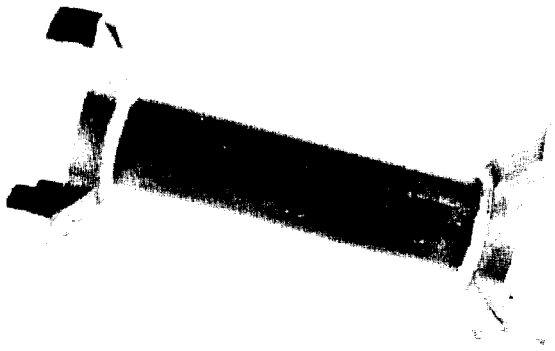


Fig. 6. Lock bar for the wheels of a highway/railroad vehicle. The cast ductile-iron lugs are welded to the steel tube with the flux-cored wire.

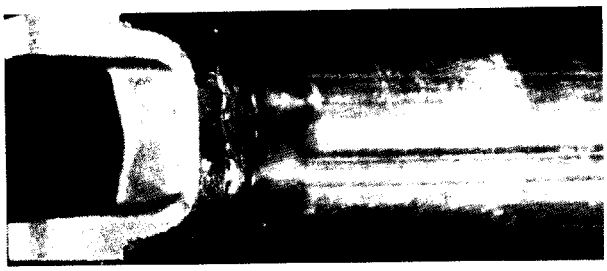


Fig. 5. Ductile iron universal-joint yoke welded to steel drive shaft with the flux-cored wire.

bar consists of two square-cornered locking lugs welded to the ends of a carbon steel tube. Cast steel has been used for the lugs because of the welding requirements. Prototype units having lugs of a ferritic ductile iron and welded with the cored wire are now being field tested. The use of ductile iron instead of steel would significantly reduce the cost of the units.

Hydraulic Cylinders

Figure 7 shows a section of a prototype hydraulic cylinder for farm machinery. The cylinder incorporates a cast head of ductile iron welded with the cored wire. The previous standard material for the head was cast steel. Substitution of ductile iron for steel required only a slight modification in joint design. Cylinders with welded cast-iron heads have met all test requirements and are scheduled to become standard production items. Cost studies project a substantial savings.

Figure 8 illustrates another hydraulic cylinder in which the cast steel head was redesigned in malleable iron and welded with the cored wire. The cylinder, which has a diameter of 3 in. (7.6 cm), is now in commercial production. The cast-iron heads are being welded with the cored wire at the same speed used for welding the cast steel with mild-steel wire. The malleable iron heads are being produced at a net savings of 55%, including cost difference for the welding products.

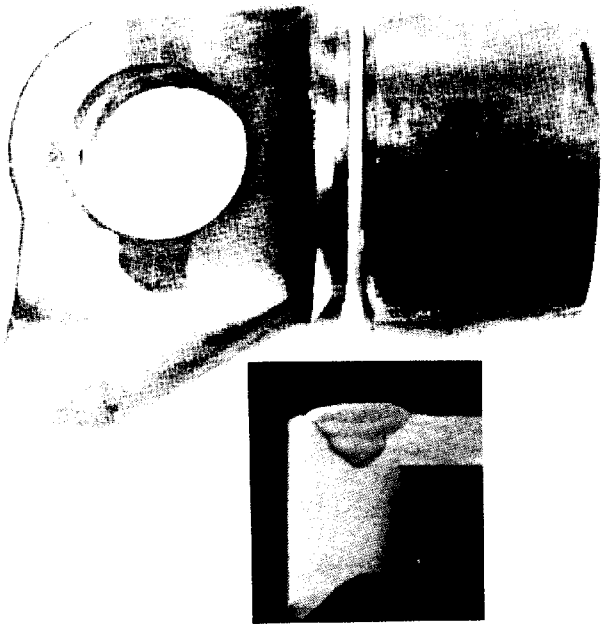


Fig. 7. Cast ductile iron head with a portion of the steel hydraulic cylinder to which it will be welded. Inset shows cross section of a finished weld bead.

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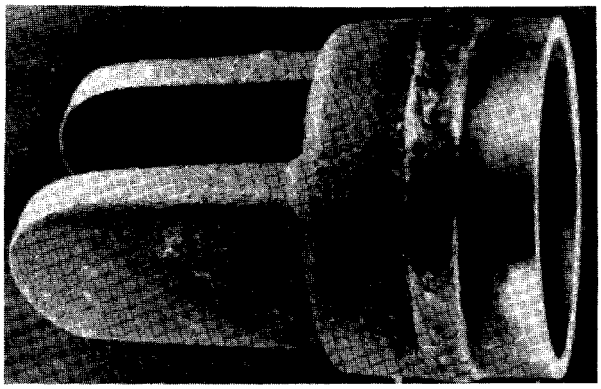


Fig. 8. Welded joint between the malleable iron head and carbon steel tube of a mass-produced hydraulic cylinder.



Fig. 9. Wheel-spindle assembly for large off-road vehicle produced by welding the steel spindle to its ductile iron housing.

Wheel Spindles

Flux-cored arc welding with the cored wire has been adopted as the standard production method for the heavy-equipment wheel spindle shown in Fig. 9. A single-pass fillet weld is used to join the steel spindle to the ductile iron backing plate into which the spindle is pressed. The welded joint meets all the requirements imposed by the vehicle's strenuous off-road service.

Other Applications

The most apparent advantage offered by the new flux-cored wire is the opportunity to reduce materials cost through the use of cast irons instead of steel. In many other instances, however, the use of the cored wire can improve existing cast-iron assemblies. In some applications, cast irons are now being considered for design reasons in addition to cost.

Figure 10 illustrates an example of the use of the cored wire to improve an existing cast-iron component. The hollow, ductile iron tail shaft for an automobile transmission requires installation of a mild-steel disc to prevent loss of transmission fluid. The disc is rolled into the end of the shaft and the resulting mechanical seal is unreliable in service. The cored wire enables the disc to be welded, providing a permanent seal.

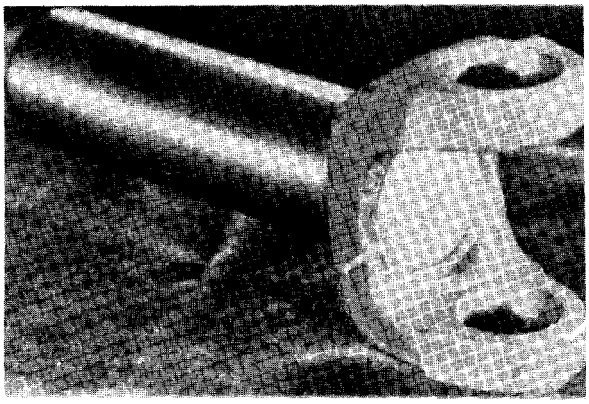


Fig. 10. Ductile iron tailshaft for automobile transmission with unwelded seal plate.

Another application for the cored wire is spot welding of various clips and brackets to cast-iron components. Resistance welding, although widely used to attach such items to steel, produces very brittle welds in cast irons. The high-Ni weld metal provides excellent ductility.

Figure 11 shows a prototype steering-column assembly for heavy off-road equipment. Standard units consist of a stamped steel flange welded to the carbon-steel steering column. Designers are now considering use of the illustrated malleable iron flange because of the ability of cast iron to dampen the engine-compartment noise transmitted to the operator by the steel flange. The machined, cast flange also provides a more consistent mating surface.

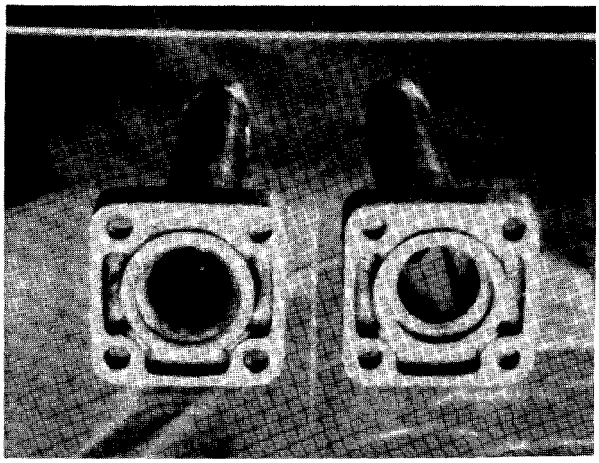


Fig. 11. Malleable iron steering-column flange for off-road equipment. The unit on the left was welded with the flux-cored wire. An unwelded joint is shown on the right.

For those instances in which manual welding is being used on the production line, substantial cost savings may be possible through substitution of automatic welding with the cored wire. Figure 12 shows an example now in commercial production. The balance weights on brake drums for large trucks were formerly attached by manual shielded-metal-arc welding. The cored wire permits the use of automatic welding to plug-weld the carbon-steel weights on the drum, greatly reducing welding costs. The high current density associated with the flux-cored wire enables fusing through the casting skin without creating hard spots on the inner drum surface.

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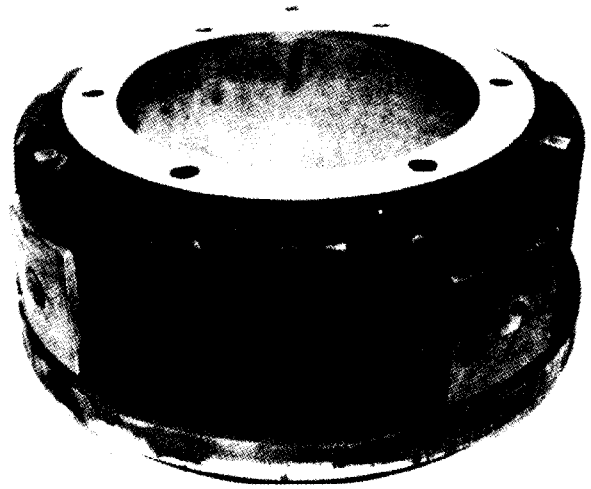


Fig. 12. Brake drum with carbon steel balance weights attached with the flux-cored plug welds.



Fig. 13. Cast iron cylinder liner with carbon-steel water jackets welded to it with the flux-cored wire.

Another example of the use of automatic welding with the cored wire to reduce assembly costs is shown in Fig. 13. Production of a large diesel engine involves joining two carbon-steel water jackets to the illustrated cast cylinder liner of a proprietary gray iron. The standard joining method has been by brazing with an expensive silver brazing alloy. Units welded with the flux-cored wire are now being tested and offer substantial cost savings.

The flux-cored wire can also be used to increase foundry production. The rejection rate for large or complex castings can be reduced by welding together smaller, easier-to-cast components. In one example, a 195-lb (88 kg) transmission case for heavy equipment has been difficult to cast in one piece because of the intricate core configuration. Studies indicated that the shape could readily be cast in two sections and welded with the flux-cored wire.

Summary

The development of a new welding product, a flux-cored wire, has made assembly-line welding of cast irons a commercial reality. High-quality welds are now being produced in ductile, malleable and gray irons by automatic, high-deposition-rate processes.

The availability of an automatic welding method for cast irons gives manufacturers of various types of assembled equipment the opportunity for dramatic reductions in materials costs. Previously, many components that required welding had to be produced in forged or cast steel. Those parts can now be redesigned in economical cast irons and welded on the assembly line at the same production rates and with equivalent joint integrities.

The new flux-cored wire is presently being used or evaluated for numerous applications in highway vehicles, construction equipment, farm machinery and other welded structures. The results of both experimental testing and in-service use clearly indicate that welded cast irons will become important materials of construction.

References

1. NI-ROD (Trademark of Huntington Alloys, Inc) FC 55 cored wire.

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Cast Irons

R. A. Bishel, J. G. Bielenberg, R. J. Dybas, K. L. Johnson, and A. B. Malizio

GENERAL DESCRIPTION

Cast iron is a common term for a series of ferrous alloys that normally contain more than 2 percent carbon and 1 to 3 percent silicon, as well as phosphorus and sulfur. Alloy cast irons may also contain one or more other elements deliberately added to provide desired properties such as strength, hardness, hardenability, or corrosion resistance. Common alloying additions are chromium, copper, molybdenum, and nickel.

In general, cast irons have lower melting ranges than those of steels, are quite fluid when molten, and undergo moderate shrinkage during solidification and cooling. The toughness and ductility of cast iron are lower than those of steel; these low properties limit the applications of cast iron.

The mechanical properties of an iron casting depend upon the type of microstructure as well as the form and distribution of the microstructural constituents. One microstructural constituent that has a significant effect is the free graphite (carbon). The amount, size, and shape of the graphite particles affect the strength and ductility of a cast iron. Consequently, cast irons may be classified by the characteristics of the graphite as it appears in a polished section.[1]

The microstructure of the matrix surrounding the graphite particles also influences the mechanical properties of a casting. The matrix is basically steel that may be ferritic, pearlitic, austenitic, or martensitic. The specific matrix in a casting will depend upon the chemical composition, cooling rate, and heat treatment of the casting.

There are four basic types of cast iron: gray iron, malleable iron, ductile or nodular iron, and white iron. Compacted graphite (GC) cast iron, a fifth type not yet standardized, has a microstructure intermediate to those of gray and ductile cast irons. Its mechanical properties fall within the range of 35 to 60 ksi yield strength (0.1 percent offset), 45 to 75 ksi tensile strength, and 1 to 6 percent elongation.

Cast irons contain more carbon than high carbon steels. They also have appreciable amounts of silicon that influence the structural form of the carbon in a casting. As a consequence, the metastable iron-iron carbide system coexists with the stable iron-graphite system in an iron casting. This duality of carbon form produces a metallic matrix in a casting that has the microstructure and attributes of steel. Additionally, the uncombined graphitic carbon is distributed throughout the matrix in a variety of geometric shapes including tiny particles, various flake-like forms, and spheroids. The size, shape, and distribution of the graphite are incorporated in casting specifications because they influence both the mechanical and the physical properties of a casting.

The matrix structure and graphite character in a particular casting are achieved by careful balance of its chemical composition, degree of inoculation, rate of solidification, and control of the cooling rate. Solidification and cooling rates are affected by the section thickness and the rate of heat transfer to the mold. Castings are heat treated to achieve properties that cannot be obtained in the as-cast condition. Annealing, austenitizing and quenching, and tempering heat treatments are used to achieve the desired microstructure in the metallic matrix.

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1. See ANSI/ASTM A 247-§7 (1978), *Standard Method for Evaluating the Microstructure of Graphite in Iron Castings.*

Source: Welding Handbook, Vol. 4, 7th Ed., 1982

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WHITE CAST IRON

White cast iron is formed when the carbon does not precipitate as graphite during solidification but remains in combination with iron, chromium, or molybdenum as carbides. This iron is hard and brittle, and has a white crystalline fracture appearance. White cast iron is normally considered unweldable because of the absence of adequate ductility to accommodate thermal stresses in the base metal.

GRAY CAST IRON

Gray cast irons are iron-carbon-silicon alloys that contain uncombined carbon in the form of graphite flakes, as shown in Fig. 5.1. This type of cast iron is named from the gray appearance of a fractured surface.

Gray iron castings for general engineering and automotive applications are divided into ten classes or grades based on minimum tensile strength, as listed in Table 5.1. Hardness, matrix microstructure, chemical composition, pressure tightness, and radiographic soundness in various combinations are sometimes specified to meet service requirements.

The tensile strength, hardness, and microstructure of a gray iron casting are influenced by several factors including chemical composition, design, the characteristics of the mold, and cooling rate during and after solidification. Castings with similar properties can be produced using different raw material mixes or various melting and casting practices, or both.

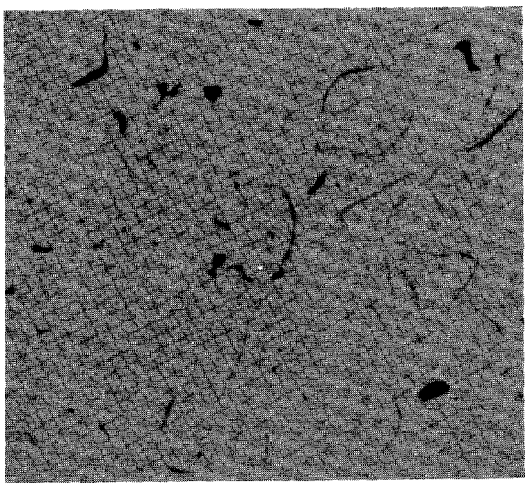


Fig. 5.1–Graphite form in gray cast iron (×250)

Copper, chromium, molybdenum, and nickel are frequently added to gray cast iron to control matrix microstructure and graphite formation. They also can improve the corrosion resistance of the cast iron to certain media.

Castings are heat-treated to obtain desired mechanical properties. Gray iron, like steel, can be hardened by quenching from a suitable temperature (1600°F), but it will be brittle. Tempering after quenching is necessary to improve toughness and decrease hardness.

Gray cast iron lacks ductility because of the flake form of the graphite. Since the graphite has virtually no strength, the numerous flakes create internal notches in the steel matrix. Fracture can initiate easily at these sites and progress rapidly across the section without plastic strain.

MALLEABLE CAST IRON

Malleable cast iron is produced by heat treating a white iron casting of suitable composition. The initial formation of white cast iron is promoted by (1) low carbon and low silicon contents, (2) the presence of carbide forming elements such as chromium, molybdenum, and vanadium, and (3) rapid solidification and cooling. The white iron casting is then heated in a controlled atmosphere furnace to a temperature above the eutectoid temperature (usually 1700°F) and held for several hours. This treatment permits the dissolved carbon in the austenite to precipitate as irregular nodules of graphite, known as *temper carbon*.

As such a casting cools slowly in a furnace through the eutectoid transformation range, the remaining excess carbon in the austenite precipitates onto the existing temper carbon nodules. The resulting graphite form is shown in Fig. 5.2. This treatment produces a ferritic matrix. A pearlitic or martensitic matrix can be obtained by alloying or by air or liquid quenching from the austenitizing temperature. These two matrix structures are normally tempered after quenching. The minimum mechanical properties of standard malleable irons are given in Table 5.2.

Malleable irons have some ductility because the temper carbon is in the form of nodules rather than flakes, as found in gray cast iron. The effect of these nodules on ductility is less severe than that of the flake graphite. The strength and ductility of malleable irons are related to the metallurgical structure of the matrix as well as the distribution of the free graphite. If the cast iron is reconverted to a metastable state by quenching

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Table 5.1
Mechanical properties of standard gray cast irons

Class or Grade	Minimum tensile strength, ksi[a]	Hardness range, HB[b]	Matrix microstructure
G1800[c]	18	187 max	Ferritic-pearlitic
20[d]	20		
G2500	25	170 to 229	Pearlitic-ferritic
25	25		
G3000	30	187 to 241	Pearlitic
30	30		
G3500	35	207 to 255	Pearlitic
35	35		
G4000	40	217 to 269	Pearlitic
40	40		
45	45		
50	50		
55	55		
60	60		

a. Minimum tensile strength of cast test bars based on the thickness of the controlling section of a casting. Refer to the appropriate specification.
b. Brinell hardness using a 10mm ball and 3000 kg load.
c. GXX00 grades refer to ANSI/ASTM A 159-77, *Specification for Automotive Gray Iron Castings.*
d. Two digit designations refer to ANSI/ASTM A48-76, *Standard Specification for Gray Iron Castings.*

from a high temperature, it must be malleabilized again to regain ductility.

DUCTILE CAST IRON

Ductile cast iron and gray cast iron are similar with respect to carbon and silicon contents, and in terms of general foundry practice for the production of castings. However, they differ in the geometric form assumed by the free graphite. In ductile cast iron, the graphite is caused to nucleate throughout the metal matrix in the form of spheroids, as shown in Fig. 5.3. As with malleable cast iron, the agglomeration of the free graphite into nodular form avoids the internal sharp notches associated with flake graphite. This accounts for the higher strength and ductility of ductile cast iron when compared to gray cast iron of similar composition. The production of ductile iron castings does not require the long-time heat treatment required to produce malleable iron castings. The properties of standard ductile cast irons are given in Table 5.3.

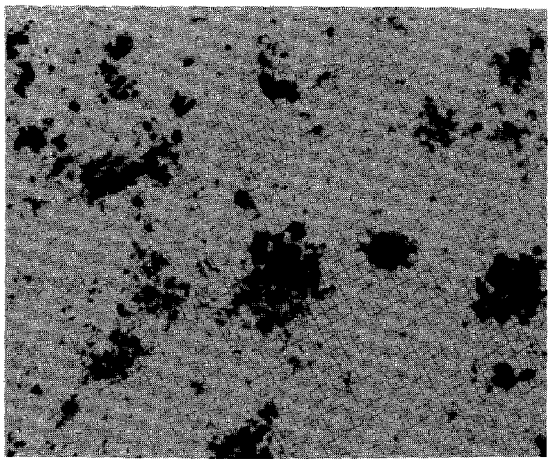


Fig. 5.2–Graphite form in malleable cast iron (× 100)

Spheroidization or nodulization of the graphite is achieved by introducing magnesium or cerium into a low-sulfur melt, preferably one containing not more than 0.02 percent sulfur. A low sulfur level is usually achieved by adding calcium oxide, calcium carbide, or sodium carbonate to the molten iron.

Magnesium can be introduced in elemental form if it is diluted or otherwise shielded to prevent instantaneous vaporization. Various proprietary alloys of magnesium combined with one or more of the elements nickel, iron, silicon, and calcium are frequently used. A residual concen-

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Table 5.2
Mechanical properties of standard malleable iron castings

Class or grade	Microstructure	Minimum tensile strength, ksi	Minimum yield strength, ksi	Minimum elongation in 2 in., percent	Hardness range, HB[a]	Heat treatment[b]
			ASTM A47-77			
32510	Ferritic	50	32.5	10		
35018	Ferritic	53	35	18		
			ASTM A220-76			
40010	Pearlitic	60	40	10	149 to 197	
45006	Pearlitic	65	45	6	156 to 207	
50005	Pearlitic	70	50	5	179 to 229	
60004	Pearlitic	80	60	4	197 to 241	
			ASTM A602-70[c]			
M4504	Ferritic-pearlitic	65	45	4	163 to 217	AQ&T
M5003	Ferritic-pearlitic	75	50	3	187 to 241	AQ&T
M5503	Martensitic	75	55	3	187 to 241	LQ&T

a. Brinell hardness using a 10 mm ball and 3000 kg load.
b. AQ&T–air quench and temper; LQ&T–liquid quench and temper.
c. Automotive castings.

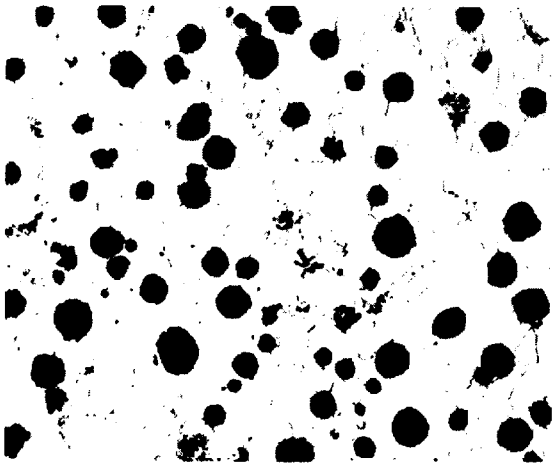


Fig. 5.3–Graphite form in ductile cast iron (×100)

tration of about 0.035 percent magnesium is necessary to produce full nodulization. If magnesium is lost from the melted zone of the base metal during welding of a ductile iron casting, the shape of the graphite along the weld interface can change. A welding process and procedure that minimize heat input and time in the molten state will minimize graphite degradation.

COMPACTED GRAPHITE CAST IRON

Compacted graphite (CG) cast iron is produced in a manner similar to ductile cast iron, but careful control of the magnesium addition to the melt is required. The graphite in this cast iron, shown in Fig. 5.4, forms in irregular geometric shapes that have a lesser surface-to-volume ratio than the flakes in gray cast iron; but it does not assume the spheroidal shape found in ductile cast iron. This intermediate form of graphite imparts a combination of desirable qualities to a casting. The mechanical properties of the three types of cast iron are compared in Table 5.4. One method for attaining the desired graphite form is to inoculate a low-sulfur melt with a magnesium ferro-silicon alloy that contains several additional elements to enhance the performance and reliability of the CG cast iron.

During solidification, CG cast iron shrinks less than ductile cast iron. It also has good thermal shock resistance. High-strength CG cast iron differs from gray cast iron in that smaller alloy

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Table 5.3
Mechanical properties of typical ductile cast irons

Grade[a]	Microstructure	Minimum tensile strength, ksi	Minimum yield strength, ksi	Minimum elongation in 2 in., percent	Heat treatment[b]
60-40-18	Ferritic	60	40	18	A
65-45-12	Ferritic-pearlitic	65	45	12	
80-55-06	Ferritic-pearlitic	80	55	6	
100-70-03	Pearlitic	100	70	3	AQ&T, LQ&T
120-90-02	Martensitic	120	90	2	LQ&T

a. Refer to ANSI/ASTM A536, Standard Specification for Ductile Iron Castings.
b. A – ferritizing anneal; AQ – air quench; LQ – liquid quench; T – temper.

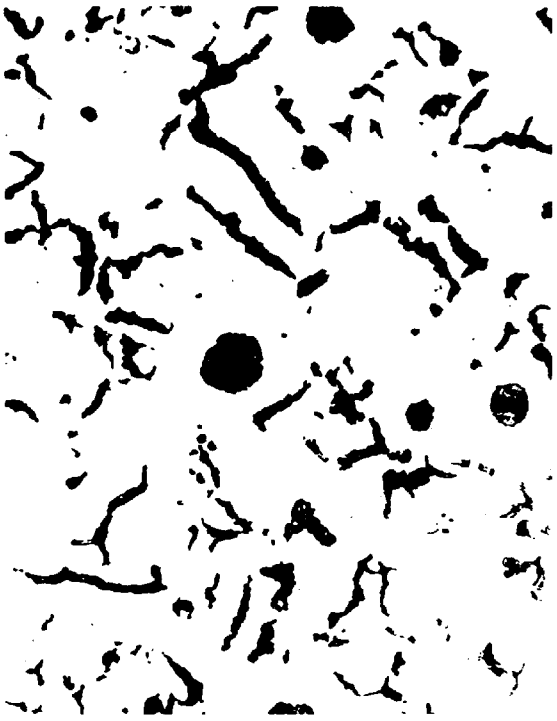


Fig. 5.4–Graphite form in compacted graphite cast iron

additions are needed to produce equivalent strength properties. The welding precautions described for ductile cast iron should also be followed with CG cast iron to minimize alteration of the graphite form.

APPLICATIONS

Each type of cast iron is used for a number of applications, many of which are covered by standards of the American National Standards Institute (ANSI), the American Society for Testing and Materials (ASTM), and the Society of Automotive Engineers. A standard sometimes contains special requirements for a cast iron to meet specific service conditions. The appropriate specification should be consulted for those requirements.

Gray Cast Iron

Gray cast iron parts are used for many applications in the machinery and automotive industries. Typical automotive parts covered by specifications are brake drums, clutch plates, and cam shafts. Furnace parts, ingot and glass molds, and melting pots that operate at elevated temperatures are made of gray cast iron, as are various types of pipe, valves, flanges, and fittings for both pressure and nonpressure applications.

Ductile Cast Iron

Some culvert, sewer, and pressure pipe, as well as fittings, valves, and pumps are cast of ductile iron to specifications. The advantages of these products are their relatively good toughness and weldability when compared to similar components of gray cast iron. Ductile cast iron piping systems are commonly fabricated by welding.

Malleable Cast Iron

Malleable cast iron is used for flanges, pipe

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Table 5.4
Comparison of the mechanical properties of gray, compacted graphite, and ductile cast irons

Type	Approx. carbon equivalent[a]	Tensile strength, ksi	Yield strength, ksi[b]	Elongation in 2 in., percent
Gray–Class 25	4.4	25	—	<1
Gray–Class 45	3.6	45	—	<1
Compacted graphite	4.2	45-75	35-60	1-6
Ductile	4.2	60-100	40-80	6-22

a. CE = %TC + 0.33 (%P + %Si), where TC is total carbon
b. 0.1% offset

fittings, and valve parts. A number of automotive parts of malleable cast iron include steering components, compressor crank shafts and hubs, transmission and differential parts, connecting rods, and universal joints. These products are covered by appropriate specifications.

WELDING CONSIDERATIONS

UTILIZATION

Welding or braze welding of cast iron is generally used for (1) the repair of foundry defects, (2) the fabrication of components, or (3) the salvage of parts that break or wear out in service. All types of cast irons except white iron are considered to be weldable but to a lesser degree than carbon steels. The best results are achieved with low-strength cast irons. Good joint efficiency is possible with established welding procedures for cast irons with tensile strengths up to and including 70 ksi. At higher strength levels, matching strength in a welded joint is increasingly difficult to achieve.

WELDABILITY

All cast irons have a common problem affecting their weldability, namely too much carbon. The manufacturing process, that is, casting and sometimes heat treating, is capable of producing microstructures that possess useful mechanical properties. However, a welding thermal cycle ordinarily does not produce a desirable microstructure. The iron immediately adjacent to the weld metal is heated to an excessive temperature, and the cooling rate of the entire heat-affected zone is too rapid. Carbides tend to form in the base metal immediately adjacent to the weld metal, and the remainder of the heat-affected zone tends to form high-carbon martensite. Both of these microstructures are very brittle and are subject to cracking, either spontaneously or during service. The degree of brittleness and propensity to cracking depend to some extent upon the type of cast iron, the condition of heat treatment, and the welding procedure.

Fusion welding, because of its localized nature, produces thermal stresses in the weld area. A base metal must be capable of some local plastic deformation to accommodate these welding stresses or cracking will likely result. Ductile and malleable cast irons having a ferritic matrix are better suited to absorb welding stresses than is gray cast iron.

As with steels, phosphorus in cast iron has an adverse effect on weldability; it should be held to less than 0.10 percent. Repair welding of high-phosphorus castings requires special techniques.

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Effects of Graphite Form

The composition and microstructure of a cast iron affect the amount of carbon in the heat-affected zone that is dissolved during welding. This, in turn, influences the brittleness and crack susceptibility of the welded joint. To minimize the formation of massive carbides and high-carbon martensite, it is most helpful to have the carbon present as spheroids that have a low surface-to-volume ratio. The smaller the surface area of the graphite in contact with an austenitic matrix, the lesser is the amount of carbon in the microstructure at room temperature. Graphite flakes in gray cast iron display the greatest tendency to dissolve in austenite because of their relatively large surface area. However, graphite in any form dissolves slowly and often remains in the weld metal. In general, melting during fusion welding is a reversal of the casting solidification process, and those areas last to solidify during casting are the first ones to melt during welding.

Cast iron typically includes the following:

3.5 percent carbon
2.5 percent silicon
0.5 percent manganese
0.04 percent phosphorus
0.06 percent sulfur

The addition of 0.07 percent of magnesium to a cast iron of such composition promotes the formation of nodular graphite. Increasing the manganese or sulfur content decreases graphitization, but higher silicon promotes it. High phosphorus content contributes to embrittlement of cast iron.

Effects of Filler Metal

The composition and mechanical properties of the filler metal are very important considerations in the welding of cast iron.[2] As each weld bead is made, filler metal and base metal, or previously deposited weld metal, are melted together. The change in composition of the filler metal as a result of mixing with base metal or previously deposited metal is called dilution.[3]

Dilution should be minimized by welding with the lowest heat input consistent with sound weld metal. The high carbon, silicon, phosphorus, and sulfur levels in cast iron can have an adverse effect on the mechanical properties and soundness of the weld metal. For example, when cast iron is welded with carbon steel filler metal, the carbon content of the deposited metal will usually increase as a result of dilution. The weld metal can be hard with low ductility and difficult to machine unless a suitable postweld heat treatment is applied. It might also be crack sensitive. Inability of the weld metal to yield and relieve welding stresses can result in cracking in the adjacent cast iron heat-affected zone.

The mechanical properties of the weld metal employed on cast iron can play a major part in the success of the operation. When the yield strength of the weld metal is low, the stresses imposed on the cast iron during cooling are relatively low. This reduces the likelihood of cracking. During service, a soft weld metal can creep and relieve stresses in the cast iron. Nickel or nickel alloy weld metal is very effective in this respect, and considerable use is made of nickel and nickel alloy filler metals for arc welding cast iron. Another advantage of these types of weld metal is the ease with which they can be machined in the as-welded condition.

PREHEATING

The formation of a hard and brittle heat-affected zone can lead to cracking during cooling or in service. Low heat input with arc welding limits the width of the heat-affected zone but a band of hard brittle iron can still form adjacent to the weld metal. Heat-affected zone hardness can be limited by preheating in combination with slow cooling after welding. Preheating provides lower cooling rates in both the weld metal and the heat-affected zone than occurs without it. A low cooling rate during and after austenitic transformation reduces the amount of martensite formed and thus the hardness.

The preheat temperature and extent of preheating depend upon the type of cast iron being welded, the mass of the casting, the welding process, and the type of filler metal. Recommended preheat and interpass temperature ranges are given in Table 5.5. In general, ferritic ductile and ferritic malleable cast irons can be arc welded with lower preheat than can the pearlitic types because ferritic cast irons have better ductility.

All cast irons generally need to be preheated when oxyacetylene welding to reduce the heat input requirements. High preheat is needed when using a cast iron filler metal because the weld

2. Refer to AWS A5.15-82, *Specification for Welding Rods and Covered Electrodes for Cast Iron*.

3. Dilution is discussed in more detail in Chapter 12.

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Table 5.5
Recommended preheat and interpass temperature range for welding cast irons

Type of cast iron	Matrix microstructure	Temperature range, °F[a]	
		Arc welding	Oxyacetylene welding
Gray	—	70 to 600	800 to 1200
Malleable	Ferritic	70 to 300	800 to 1200
Malleable	Pearlitic	70 to 600	800 to 1200
Ductile	Ferritic	70 to 300	400 to 1200
Ductile	Pearlitic	70 to 600	400 to 1200

a. A suitable preheat temperature for a particular application depends upon the welding process, filler metal, mass of the casting, welding process, and postweld heat treatment.

metal has low ductility near room temperature. A filler metal that deposits relatively low strength, ductile weld metal, such as ENi-CI or ENi-CI-A electrodes, can be used with the base metal at or slightly above room temperature. The weld metal can readily yield during cooling and relieve welding stresses that might otherwise cause cracking in the weldment.

Cracking from unequal expansion can take place during the preheating of complex castings or when the preheating is confined to a small area of a large casting. Local preheating should be gradual. Preheating of either a large section of the casting where the welding is to be done or the entire casting in a uniform manner is recommended. In any case, the preheating temperature should be maintained during welding, and welding should be completed before the casting is cooled slowly to room temperature.

Where possible, the preheating pattern should be designed to place the welded joint in compression after cooling to room temperature. The part configuration should be thoroughly analyzed before preheating to minimize the introduction of thermal stresses from this operation that might cause subsequent failure.

To ensure proper control, preheat and interpass temperatures should be monitored. Contact pyrometers, temperature-sensitive crayons, or thermocouples are means of monitoring temperature.

SURFACE PREPARATION

All casting skin and foreign material should be removed from the joint surfaces and adjacent areas. Castings that have been in service are often impregnated with oil or grease that can be removed by solvents or steam cleaning. Where possible, the casting should be heated uniformly at about 700°F for 30 minutes, or for a shorter time at 1000°F, using an oxyfuel gas torch or circulating air furnace. An alternative is to heat the weld area by depositing the first weld pass and then removing it by a suitable means because the weld metal is generally very porous. This welding and removal operation is repeated until the weld metal is sound. Then, the weld is completed in a normal fashion. Castings that have been impregnated with a plastic or glass sealer should not be repair welded in any case because the sealer will produce gross porosity or inhibit bonding.

Brazing or soldering filler metals will not wet exposed graphitic carbon in cast iron and will result in a poor metallurgical bond. Where wetting difficulties are encountered, several cleaning methods can be used. One method is a proprietary electrochemical cleaning operation that produces a surface essentially free of graphite, sand, silicon, oxides, and other contaminants. The process uses a catalyzed molten salt bath operating at 850° to 950°F in a steel tank. Direct current is passed through the bath with the work as one electrode and the steel tank as the other. The direction of current flow is occasionally reversed to produce alternate reducing and oxidizing effects. A water rinse completes the surface treatment.

Abrasive blasting with steel shot may be a suitable method for preparing the surfaces of ductile and malleable cast irons but not for gray cast irons. Searing the surface with an oxidizing flame or heating the casting to about 1650°F in a strongly decarburizing atmosphere may be suitable for some applications.

Before any cleaning procedure is used in production, wetting tests should be conducted

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using the proposed filler metal and joining procedures. The filler metal should be applied to a cleaned, flat surface and then examined visually for good wetting. If the surface is not uniformly wetted, it has not been sufficiently cleaned.

POSTWELD HEAT TREATMENT

Thermal stress relieving is desirable for fully restrained welds, welds intended for use in severe service, and welds subject to close machining tolerances. Normally, stress relieving is done immediately after welding by increasing the temperature of the entire casting into the 1100° to 1150°F range. The casting is held at temperature for about one hour per inch of thickness. The cooling rate should not exceed 50°F per hour until the casting has cooled down to about 700°F. Some reduction in hardness can be achieved by this stress relieving treatment but it will be slight because the carbides are stable at 1100°F.

Heat treatment at 1650°F followed by slow cooling to 1000°F or lower results in maximum softening and stress relief. To obtain optimum ductility, the weldment should be given this heat treatment immediately after welding. Reduction in strength must always be considered when selecting a stress-relieving temperature.

For the best results with ductile cast iron, the welded casting should be immediately placed into a hot furnace (1100° to 1200°F) and the temperature raised to 1650°F. The casting should be held at temperature for two to four hours. It is then cooled to 1300°F, held there for five hours, and then cooled down to 1100°F in the furnace. The casting may then be cooled to room temperature in the furnace or in still air.

If a casting is welded with a high (1200°F) preheat and interpass temperature, it should be slowly cooled by covering it with an insulating blanket, vermiculite, or hot sand. Malleable iron castings may be reheat-treated after welding.

FILLER METAL SELECTION

Several factors must be considered when choosing a filler metal for welding cast iron. The important ones are as follows:

(1) The type of cast iron
(2) Mechanical properties desired in the welded joint
(3) Tolerance of the filler metal to dilution by the base metal
(4) Ability of the weld metal to yield and relieve welding stresses
(5) Machineability of the weld zone
(6) Color matching
(7) Applicable welding process
(8) Cost

Filler metals are available for welding cast iron by the shielded metal arc, flux cored arc, gas metal arc, and oxyacetylene welding processes. In addition, filler metals are available for braze welding. In special cases, cast irons may be welded using gas tungsten arc welding and a suitable filler rod. Typical filler metals are given in Table 5.6. Other suitable proprietary filler metals are also available.

ARC WELDING

JOINT DESIGN

Joint designs commonly used for arc welding of carbon steels are generally suitable for welding cast iron sections together. The root opening should be wide enough to permit good fusion into the root faces and into a backing plate, if used. When practical, thick sections should be welded from both sides using either a double-V- or double-U-groove preparation.

When repairing cracked castings, a hole about 0.13 inch or larger in diameter should be drilled at the end of each crack to prevent further propagation. Then, sufficient cast iron should be removed to eliminate the crack and provide room to properly manipulate a welding electrode or torch during repair welding.

SHIELDED METAL ARC WELDING

Cast irons can be arc welded with nickel, nickel alloy, mild steel, and copper alloy covered electrodes. Bare gray cast iron electrodes are used for special applications. Filler metal selection depends upon the type of cast iron to be joined and the application. In any case, dilution

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with the cast iron should be kept to a minimum. Covered electrodes for cast iron welding are listed in Table 5.6. Figure 5.5 shows the repair welding of a casting with a covered electrode.

Nickel Alloy Electrodes

The nickel alloy electrodes, such as ENi-CI and ENiFe-CI, are specifically designed for the welding of cast irons. Deposited metal from each type of electrode has a carbon content well above the solubility limit. The excess carbon is rejected as graphite during solidification of the weld metal. This reaction causes an increase in volume that tends to minimize weld metal shrinkage during solidification. This, in turn, reduces residual stresses in the weld metal and the cast iron heat-affected zone.

The nickel (ENi-CI) electrode produces a softer weld metal than the nickel-iron (ENiFe-CI) electrode. This is important where machinability of the weld metal is a factor. The ENiFe-CI weld metal has good machinability also, particularly with multiple-pass welds. In general, this electrode is a better choice because of the higher strength and ductility of the weld metal. Also, this weld metal has a greater tolerance for phosphorus from the cast iron and better resistance to hot cracking. Castings that offer a small

Table 5.6
Filler metals for welding or braze welding cast iron

Description	Form	Applicable process[a]	AWS Specification	AWS Classification
		Cast iron		
Gray iron	Welding rod	OAW	A5.15	RCI
Gray iron	Bare electrode	BMAW	–	–
Alloy gray iron	Welding rod	OAW	A5.15	RCI-A
Ductile iron	Welding rod	OAW	A5.15	RCI-B
		Steel		
Carbon steel	Covered electrode	SMAW	A5.15	ESt
Carbon steel	Covered electrode	SMAW	A5.1	E7018
Carbon steel	Bare electrode	GMAW	A5.18	E70S-2
		Nickel alloys		
93% Ni	Bare electrode	GMAW	A5.14	ERNi-1 ENi-CI
95% Ni	Covered electrode	SMAW	A5.15	ENi-CI-A ENiFe-CI
53 Ni-45 FE	Covered electrode	SMAW	A5.15	ENiFe-CI-A
53 Ni-45 Fe	Flux cored electrode	FCAW	–	–
55 Ni-40 Cu-4 Fe	Covered electrode	SMAW	A5.15	ENiCu-A
65 Ni-30 Cu-4 Fe	Covered electrode	SMAW	A5.15	ENiCu-B
		Copper alloys		
Low fuming brass	Welding rod	OAW	A5.27	RCuZn-B
Low fuming brass	Welding rod	OAW	A5.27	RCuZn-C
Nickel brass	Welding rod	OAW	A5.27	RBCuZn-D
Copper-tin	Covered electrode	SMAW	A5.6	ECuSn-A
Copper-tin	Bare electrode	GMAW	A5.7	ERCuSn-A
Copper-aluminum	Covered electrode	SMAW	A5.6	ECuAl-A2
Copper-aluminum	Bare electrode	GMAW	A5.7	ERCuAl-A2

a. OAW–oxyacetylene welding
BMAW–bare metal arc welding
SMAW–shielded metal arc welding
GMAW–gas metal arc welding

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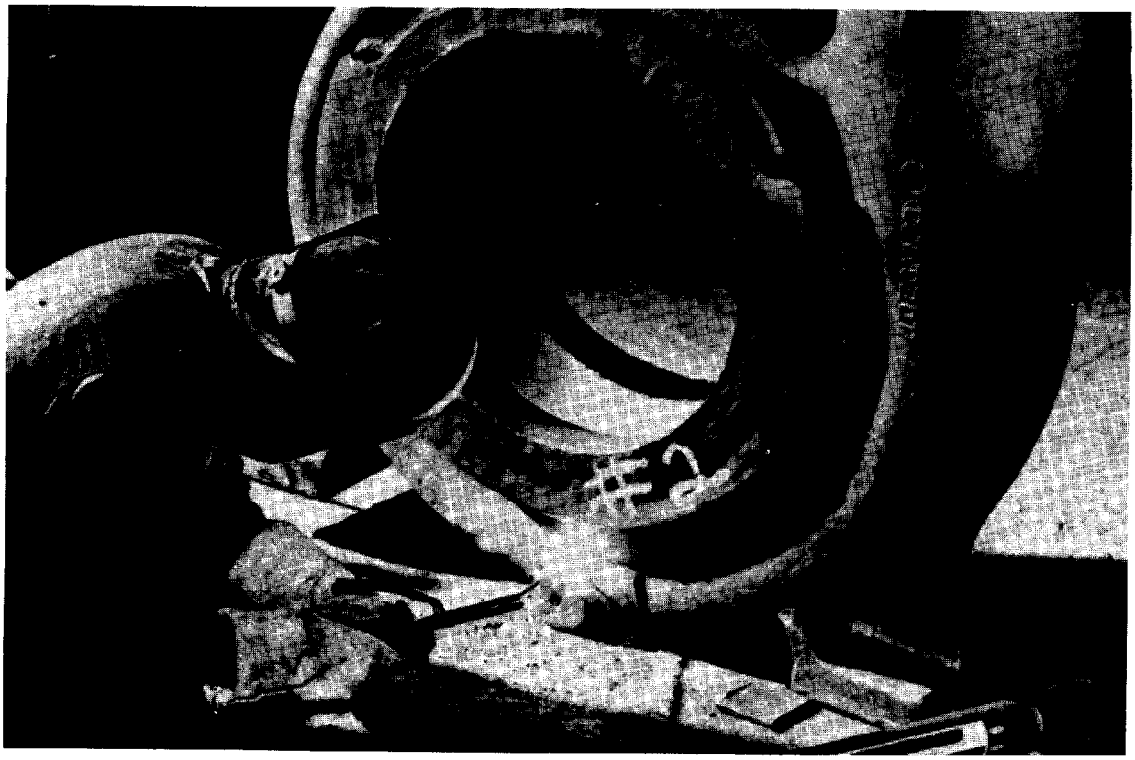


Fig. 5.5–Repairing a pump casting by shielded metal arc welding

amount of restraint can be welded with either of the nickel alloy electrodes. When welding must be done under high restraint, the ENiFe-CI electrode is preferred. This electrode is also suitable for welding cast iron to mild steel, stainless steels, and nickel alloys.

For V-groove welds, a 60- to 80-degree groove angle is suitable. For thick sections, a U-groove with 20 to 25 degree groove angle and 0.19 to 0.25 in. root radius should be used.

Welding current for a particular electrode size should be within the range recommended by the manufacturer but as low as possible, consistent with smooth operation, desired bead contour, and good fusion. When used in other than the flat and horizontal positions, the welding current should be reduced about 25 percent for vertical welding and about 15 percent for overhead welding.

The electrode should be manipulated so that the bead width is no greater than three times the nominal diameter of the electrode. If a larger cavity must be filled, the sides should be buttered with weld metal first, and then the cavity gradually filled toward the center of the repaired area. With large castings, a backstep sequence provides a more even thermal distribution.

Use of preheat is not always necessary, but it is often used. It is especially helpful in overcoming the differential mass effect encountered when welding thick to thin sections. Preheat is also beneficial for applications requiring leak tightness because it reduces the likelihood of cracking in the weld or heat-affected zone.

Peening of the hot weld bead helps to reduce welding stresses and maintain dimensions. It should be done with repeated moderate blows using a round-nose tool and sufficient force to deform the weld metal but without rupturing it. A post-weld heat treatment is sometimes used to improve the machinability of the weld heat-affected zone.

The nickel-copper-iron covered electrodes (ENiCu-A and ENiCu-B) are used to weld cast iron in a manner similar to that for nickel (ENi-CI) and nickel-iron (ENiFe-CI) electrodes. The rejection of carbon as graphite in the weld metal is also similar. However, nickel-copper-iron weld metal is more susceptible to cracking from dilution by cast iron. Therefore, dilution must necessarily be limited using appropriate welding techniques.

Mild Steel Electrodes

Mild steel electrodes, such as E7018 and

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ESt types, are used primarily to repair small cosmetic casting defects where color match is desirable and machining is not of major concern. Dilution raises the carbon content of the steel weld metal and that, in turn, increases hardenability. Therefore, the welding procedure should be designed to minimize both the dilution and the cooling rate to keep the hardness of the weld as low as possible.

Mild steel electrodes can be acceptable for joining ferritic ductile or malleable iron castings to mild steel components. However, the steel weld metal shrinks more than the cast iron and causes stresses to develop at the weld interface. The stresses can be severe enough to cause cracking in the cast iron heat-affected zone. Use of steel electrodes should be restricted to applications where the joint is not loaded in tension or in bending. In any case, a suitable preheat should be used, and the welding procedures should be thoroughly evaluated prior to production use.

Gray Iron Electrodes

Bare metal arc welding with gray cast iron electrodes has limited use for the repair welding of gray iron castings. The process is faster than oxyacetylene welding, and a machineable weld can be produced with special procedures.

The casting must be preheated to prevent cracking. During welding, the joint faces should be thoroughly melted and fused with the deposited metal. The arc should not be broken abruptly, but the arc length should be gradually increased and held for a short time to permit the crater to solidify slowly and thus avoid cracking.

For machineable deposits, the base metal should be preheated as for oxyfuel gas welding. Welding is done in the flat position with a large electrode and high amperage to produce a large, fluid weld pool. Carbon dams may be used to support the weld pool at the root and edges of the repair. Undercut is a problem with this technique, and undercut areas can be filled using oxyacetylene welding procedures. The cooling rate to room temperature should be controlled to avoid the formation of martensite in the weld zone.

Copper Alloy Electrodes

Cast iron can be arc (braze) welded with copper alloy electrodes.[4] Copper alloy weld metal has appreciable amounts of alpha phase which is soft and ductile when hot. Yielding of the soft weld metal, during cooling, limits welding stresses and reduces the likelihood of cracking. A large part of the contraction strain takes place before the cooling weld metal reaches 500°F, and it is relieved by harmless plastic stretching of the weld metal. The strength of the weld metal increases rapidly as the temperature drops to ambient with little change in ductility.

Two types of copper alloy electrodes are listed in Table 5.6. The strength of copper-aluminum weld metal will be about twice that of copper-tin weld metal. Two advantages of arc welding over oxyacetylene welding are higher welding speed and lower distortion. Also, the heat-affected zone will be narrower with less likelihood of cracking.

The welding groove face area should be large to provide a large brazed area and adequate joint strength. A V-groove with a 90- to 120-degree included angle is recommended. The joint faces should be cleaned, as described previously, to remove any graphite that can inhibit wetting.

A preheat suitable for the type of cast iron should be used. The copper alloy filler metal is deposited in stringer beads using a heat input just sufficient to obtain good wetting of the cast iron joint faces but with minimum dilution. Where possible, the arc should be directed at a previously deposited bead, and never at a corner or edge on the cast iron. The joint faces can be buttered with filler metal prior to welding the joint. The weldment should be slowly cooled to room temperature.

GAS METAL ARC WELDING

Gas metal arc welding with short-circuiting transfer is suitable for joining ductile iron. Because of the relatively low heat input with this process, the hard portion of the heat-affected zone is usually confined to a thin layer next to the weld metal. As a result, the strength and ductility of the welded joint are about the same as those of the base metal.

Bare electrodes, similar to covered electrodes for shielded metal arc welding, are recommended[5] (see Table 5.6). Typical electrodes

4. See *Welding Handbook*, Vol. 2, 7th Ed., Ch. 11 for information on braze welding.

5. See Specifications AWS A5.14-76 and AWS A5.18-79 for information on nickel alloy and carbon steel bare electrodes respectively.

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are of nickel (ENi-1) and carbon steel (E70S-2). The ERNi-1 nickel electrode is low in carbon. Therefore, deposited metal from this electrode contracts more on cooling than deposited metal from the ENi-CI covered electrode. A shielding gas appropriate for the electrode should be used.

FLUX CORED ARC WELDING

Flux cored arc welding can be used with cast irons. A flux cored electrode is available that produces weld metal with a composition and microstructure similar to that deposited by ENiFe-CI covered electrodes. The principle of graphite precipitation and the associated volume increase is the same for both electrodes.

The flux cored electrode is a self-shielding design, but it may be used with carbon dioxide (CO_2) shielding or as a submerged arc welding electrode. The electrode producer should be consulted for recommendations on usage and procedures for specific applications.

BUTTERING

A suitable procedure for arc welding cast iron uses a buttering technique that provides good weld joint ductility without a postweld heat treatment. The object of the buttering is to place the heat-affected zone of the welded joint in the buttering layer rather than in the cast iron. A layer of weld metal about 0.3-in. thick is deposited on the joint faces, and then the components are annealed immediately.

The filler metal employed for buttering can be, but is not necessarily, the same one that is subsequently used for joining the cast pieces together. However, it must have properties commensurate with the base metal after heat treatment and welding. This approach has been successfully demonstrated by shielded metal arc welding with nickel-iron (ENiFe-CI) and mild steel (E7016) covered electrodes.

OXYACETYLENE WELDING

Oxyacetylene welding procedures require large amounts of heat input during both the preheating and the welding operations. The extensive heating is a limiting factor in the application of this process to finished or semifinished castings where distortion or dimensional stability can be a problem. On the other hand, the resulting slow cooling rate lessens the tendency for brittleness in the weld heat-affected zone. Only limited success has been achieved in welding malleable cast irons by this process.

JOINT DESIGN

The joint design is normally a double-V-groove with a 90-degree groove angle. If welding can be done only from one side, the groove angle for a single-V-groove may be increased to 120 degrees. The joint groove must have sufficient width to permit good fusion of the root faces and uninterrupted torch manipulation. If a deep hole is being repaired, the groove should be elongated to avoid torch backfire.

FILLER METALS

For gray iron castings, gray iron welding rods of AWS Type RCI or RCI-A are used. The RCI-A rods contain small amounts of molybdenum and nickel, and the deposited metal is of higher strength than that of RCI rods. The melting temperature is slightly higher, the molten metal is more fluid, and welding can be done faster.

Type RCI-B rods are designed for welding ductile cast irons but can also be used with gray cast irons. These rods contain a small amount of cerium that tends to agglomerate the graphite in the weld metal during solidification and produce ductile weld deposits.

Other elements are added to the filler metals to provide desirable properties. Phosphorus additions improve the fluidity of the molten weld metal, while low phosphorus reduces the fluidity for surfacing operations. Small additions of chromium and vanadium improve strength. Cast iron weld deposits have excellent color match with the base metals.

FLUX

A flux is required during oxyacetylene welding to increase the fluidity of the fusible iron silicate slag that forms in the molten weld metal.

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This slag is difficult to remove when a flux is not used. Prefluxed welding rods are commercially available, but the more common practice is to dip the heated end of a bare welding rod into the flux and transfer it as required to the weld. Excessive flux can result in harmful slag entrapment.

Fluxes for welding gray cast iron contain borates or boric acid, soda ash, and other compounds such as iron oxide, sodium chloride, and ammonium sulfate. Most fluxes are of proprietary compositions, but a typical flux contains equal parts of boric acid and soda ash, 2 percent ammonium sulfate, and 15 percent powdered iron. Fluxes for welding ductile cast irons are similar to those used for gray cast irons except for ingredients that produce a lower melting slag.

WELDING PROCEDURE

A preheat temperature between 1100° and 1200°F is normally employed to compensate for the low heat input of the process. Thin sections can be welded with a lower preheat temperature because the heat loss into the base metal will be less than with thick sections.

Welding is performed with a neutral or slightly reducing flame using welding tips of medium or high flame velocity. The backhand welding technique is recommended. The tip of the welding rod should be kept in the molten weld pool to minimize slag inclusions and porosity in the weld metal.

The weld metal should be deposited in layers less than 1/8-in. thick. For good fusion with the joint faces, the molten pool must be kept small and the torch must be directed on the base metal. Moving the welding rod about in the molten weld pool helps to float the slag to the surface.

For ductile cast irons, melting of the base metal should be restricted to that required for good fusion. An interpass temperature of between 1100° and 1200°F should be maintained. If the temperature falls below this range, porosity may result from gas evolved by the molten base metal. Immediately after completion of welding, the casting should be insulated to provide slow cooling. Stress relieving or annealing of the casting is recommended.

BRAZE WELDING

Braze welding with an oxyacetylene torch can be used to make field repairs to castings, but it is not normally used to repair new castings because of poor color match.[6] A typical operation is show in Fig. 5.6. Copper alloy filler metals commonly used for braze welding are listed in Table 5.6. Joint strengths equivalent to fusion welds are possible with gray cast iron but not with ductile or malleable cast iron. With ductile cast iron, for example, a joint efficiency of only about 80 percent can be expected with RBCuZn-D filler metal and even less with the other RBCuZn filler metals. Corrosion resistance of the joints is generally poor.

V-groove joint designs similar to those used for arc or oxyacetylene welding are suitable. The joint faces and adjacent surfaces should be prepared as described for brazing.

The copper-zinc welding rods are used with a neutral or slightly oxidizing flame. To prevent excessive oxidation, the molten weld pool should be kept covered with a thin oxide film. Preheating with the welding torch is satisfactory for small castings. Large castings require preheating to a temperature in the range of 600° to 750°F, or higher, prior to welding. Generally, a suitable flux is applied to the welding rod by warming one end with the flame and then dipping it into the powdered flux. The flux-coated rod is applied to the joint during welding. Flux-coated welding rods are commercially available. With thick sections, flux is sometimes applied to the preheated joint faces before welding begins.

The behavior of the filler metal on the cast iron indicates when the proper temperature has been reached. If the cast iron is too cold, the filler metal will not wet and spread over the surface. If the temperature of the iron is too high, the filler metal will ball up on the surface of the cast iron. The fact that braze welding is done at a temperature of several hundred degrees below the melting point of cast iron is an advantage with respect to dilution.

6. Refer to the *Welding Handbook*, Vol. 2, Ch. 11, for additional information on braze welding.

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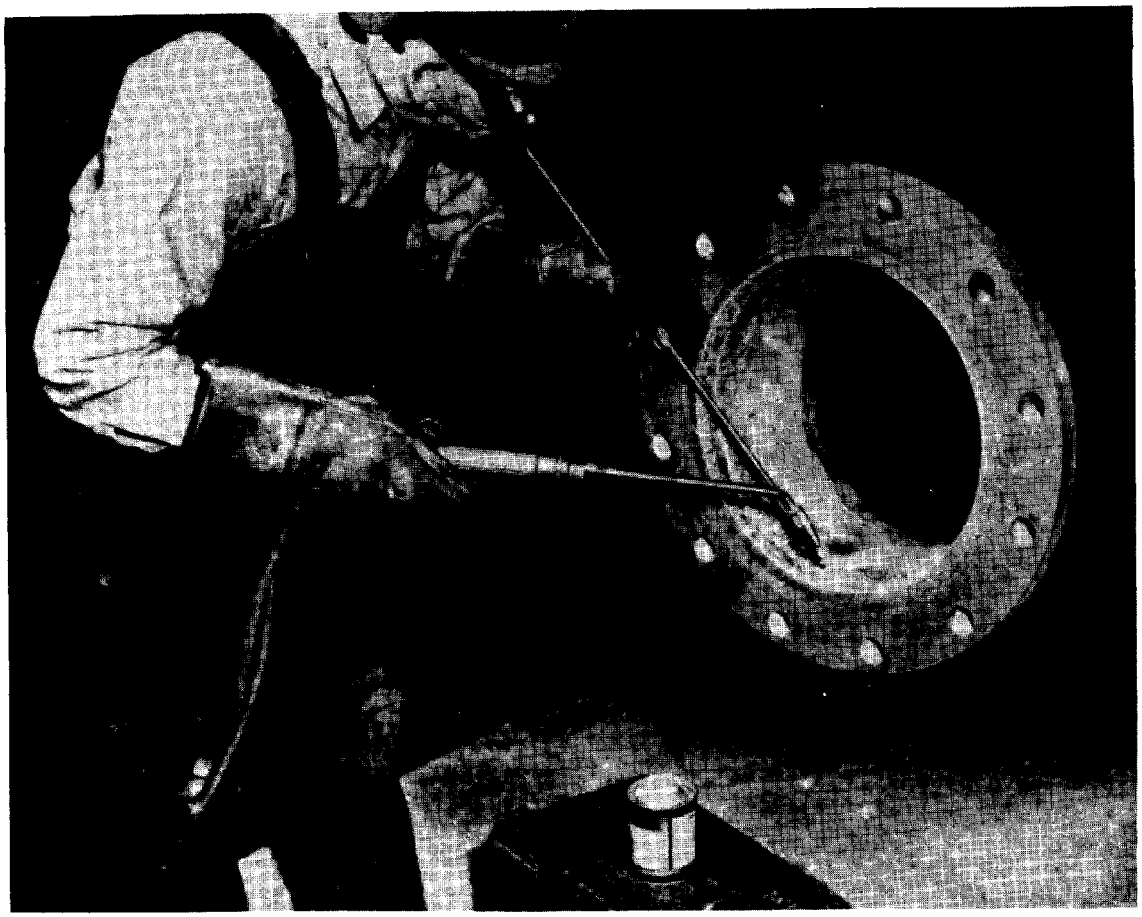


Fig. 5.6–Braze welding a cast iron flange to a cast iron pipe with an oxyacetylene torch

Welding is best done in the flat position. As welding progresses, the welding groove faces must be tinned with filler metal ahead of the weld to insure good wetting and fusion. When tinning is properly done, the filler metal penetrates the grain boundaries of the iron at the surface. After the groove faces are tinned, the weld metal is built up in layers until the joint is filled. The casting is then covered with insulation and allowed to cool slowly to room temperature.

BRAZING

FILLER METALS AND FLUXES

With proper surface preparation, any filler metal suitable for brazing carbon steel can be used for cast iron. The best suited ones are the lower melting silver brazing filler metals. Those containing nickel, such as Types BAg-3 and BAg-4, have good wettability on clean cast iron and, therefore, produce higher strength joints. Copper and copper-zinc filler metals can be used, but their high brazing temperatures must be considered. Filler metals containing phosphorus (BCuP) are not suitable for joining cast iron because the formation of brittle iron phosphide embrittles the brazed joint. Types 3A and 3B brazing fluxes are used with BAg brazing filler metals.

BRAZING PROCESSES

Any brazing processes suitable for steel are applicable to cast irons. The choice of process

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depends upon the metals being joined, the brazing filler metal used, the design of the joint, and the relative masses of the parts. Those processes that use automatic temperature control are most desirable because the cast iron must not be overheated during the brazing operation.

JOINT CLEARANCE

Joint clearance should be determined for a specific application after considering such factors as the joint design, the thermal expansion coefficients of the metals being joined, the method of heating, and the filler metal. The recommended joint clearance for BAg filler metals is 0.002 to 0.005 in. The proper clearance is easily maintained with small components, but some sacrifice must usually be made with large parts.

BRAZING PROCEDURES

In general, the handling of cast iron parts during brazing is the same as for other base metals, assuming adequate preparation of the faying surfaces as described previously. Because cast irons expand significantly with temperature and conduct heat quite poorly, the heating and cooling cycles for brazing should be designed to minimize thermal stresses. The casting should be cooled slowly from the brazing temperature.

POSTBRAZE OPERATIONS

Postbraze operations include the removal of excess flux. Warm water is usually adequate for dissolving the flux normally used with silver base filler metals. Inspection operations are then made on the cleaned joints.

OTHER JOINING PROCESSES

SOLDERING

Soldering is used to a limited extent for repair of small surface defects in iron castings. Such repairs can provide both water- and air-tight seals. Torch heating is usually employed with a slightly reducing flame. Soldering temperatures in the range of 450°-500°F are normal. The faying surfaces must be prepared as described previously. If tinning difficulties are experienced, a solder relatively high in tin (50 percent or more) might help the operation. A typical solder used to repair gray cast iron has the composition 35Sn-30Pb-35Zn.

Fluxes similar to those for soldering steel are normally used to assist wetting for a well-tinned surface. The joint should be cooled slowly to prevent tearing of the solder filler metal.

THERMIT WELDING

Iron castings can be repaired by Thermit welding, especially large structures such as machine bases or frames.[7] The shrinkage of weld metal produced by the alumino thermic reaction is significantly greater than that of gray or ductile cast iron. For this reason, difficulties can arise when the process is used for the repair of cracks that are longer than eight times the section thickness or that do not extend through the section.

The process is essentially the same as that used for the welding of steel. Cast iron Thermit mixture normally consists of aluminum powder and iron oxide to which 3 percent ferrosilicon and 20 percent mild steel punchings are added. Because of the insulation and slow cooling provided by the sand mold surrounding the joint, the weld metal is generally somewhat harder and tougher than ordinary gray cast iron but is machinable. Stress relieving is usually not necessary unless a condition of high restraint is encountered.

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7. Refer to the *Welding Handbook*, Vol. 3, Ch. 13, for more information on Thermit welding.

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CUTTING

OXYFUEL GAS CUTTING

The high carbon content of cast iron hinders oxyfuel gas cutting operations.[8] The large amount of graphite and iron carbide in the metal interferes with the oxidation of the iron matrix. Poor fluidity of the molten metal and slag prevents clearing the kerf. High quality production cuts, typical with steels, cannot be obtained. However, cast iron can be severed using special techniques, and it is advisable to preheat the casting to the temperature recommended for oxyacetylene welding.

In general, a reducing preheat flame is used to cause burning in the kerf to maintain preheat. After the starting edge is preheated through the full thickness, the cutting torch is directed backward at 45 degrees to the work surface to start the cut. After cutting is underway, the torch angle with the work surface is gradually increased to about 75 degrees or more. During cutting, the cutting torch is oscillated in a semicircular pattern transverse to the direction of travel. Cast irons are sometimes cut using a waster plate, metal powder, or chemical flux cutting techniques.

8. See the *Welding Handbook*, Vol. 2, Ch. 13, for additional information on oxyfuel gas and arc cutting methods.

ARC CUTTING

Cast irons can be cut by the air carbon arc and plasma arc cutting processes. The equipment manufacturer should be consulted for specific information and recommendations.

Metric Conversion Factors

1 ksi = 6.89 MPa
1 in. = 25.4 mm
$t_C = 0.56\ (t_F - 32)$

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NOTE: Translation of some foreign references may be found in *Welding Research Abroad* published by the Welding Research Council, 345 East 47th Street, New York, NY 10017.

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